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PART 3

DEVELOPMENT OF TITANIUM-BASE ALLOYS FOR ELEVATED TEMPERATURE APPLICATION

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MATERIALS LABORATORY
CONTRACT No. AF 33(616)-2853
PROJECT No. 7351
TASK No. 73510

WRIGHT AIR DEVELOPMENT CENTER
AIR RESEARCH AND DEVELOPMENT COMMAND
UNITED STATES AIR FORCE
WRIGHT-PATTERSON AIR FORCE BASE, OHIO

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FOREWORD

This report was prepared by the Armour Research Foundation under USAF Contract No. AF 33(616)-2853. This contract was initiated under Project No. 7351, "Metallic Materials", Task No. 73510, "Titanium Metal and Alloys", formerly RDO No. 615-11, "Titanium Metal and Alloys", and was administered under the direction of the Materials Laboratory, Directorate of Research, Wright Air Development Center, with Lt. D. A. Wruck acting as project engineer.

This report covers period of work from January 7 to December 31, 1955.

ABSTRACT

The principal objective of the work reported herein was a determination of the effects on mechanical properties of complexing the α and β phases of a promising α-β type alloy, Ti-6Al-3Mo. Tin and zirconium were employed as α complexers and chromium, manganese, and vanadium were employed as β complexers.

α complexing was found to improve creep resistance and rupture strength, while β complexing reduced these parameters below the leves of the base composition.

In addition, age hardening characteristics of Ti-Al-Ag alloys were determined. Limited creep-rupture data indicated inferiority to a Ti-6Al binary composition.

Further studies on the nature of embrittlement in binary Ti-Al alloys were carried out and results of these studies are reported.

PUBLICATION REVIEW

This report has been reviewed and is approved.

FOR THE COMMANDER:

M. R. WHITMORE
Technical Director
Materials Laboratory
Directorate of Research

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I. INTRODUCTION

This report summarizes the work performed during the period 7 January through 31 December, 1955 on the program entitled "Development of Titanium-Base Alloys for Elevated Temperature Application", under Contract No. AF 33(616)-2853.

Objectives of Program

The main objective of this program was to determine the effects on mechanical properties of complexing the α and β phases in a promising α-β type alloy. Previous research encompassing the evaluation of binary, ternary and quaternary titanium-base alloys demonstrated that Ti-7Al-3Mo has the best combination of ambient and elevated temperature mechanical properties.(1,2,3,4) Because the immediate objective of the current work was not to develop an alloy with properties superior to the 7Al-3Mo alloy but rather to evaluate the influence of complexing per se, Ti-6Al-3Mo was selected as the base material.

Tin and zirconium were selected as α phase complexing additives and the resultant alloys, namely, 6Al-3Mo-4Sn, 6Al-3Mo-4Zr and 6Al-3Mo-2Sn-2Zr, were investigated as potential forging materials.

Additions of β stabilizing elements were made by partial replacement of the molybdenum of the base material with chromium, manganese or vanadium. The total amount of β stabilizing additions was maintained at 3%. These alloys, also studied as forging materials, were the following: 6Al-2Mo-1Cr, 6Al-2Mo-1Mn, 6Al-2Mo-1V, 6Al-1Mo-1Mn-1Cr, and 6Al-1Mo-1Mn-1V.

Because of the excellent elevated temperature performance experienced with alloys based on dispersion-hardening type systems, an α-base alloy having a fully developed eutectoid microconstituent as second phase, Ti-6Al-3Cu, was examined for applicability as a forging alloy.

The 6Al-4V α + β alloy developed concurrently under Contract No. AF 33(038)-22806 and Contract No. DA-11-022-ORD-244 (Watertown Arsenal) had shown exceptional promise in the laboratory as a room and moderately elevated temperature sheet material. However, its creep-rupture properties at temperature were comparable only to the binary 6Al α alloy. As a consequence, compositional variation studies were made through substitution of molybdenum for part of the vanadium in hopes of improving creep-rupture characteristics. Modifications of the basic 6Al-4V alloy were the following: 6Al-2V-1Mo, 6Al-2V-2Mo and 6Al-3V-1Mo. Alloys were studied for either forging or sheet application and were evaluated by tensile tests at room temperature, 620°, 800° and 1020°F; creep and creep-rupture tests at 800° and 1020°F; and stability tests under stress at temperatures employed in creep testing.

Previous work on the 8Al α alloy had shown that the ductility of the alloy hinged upon the annealing treatment given it.(5) Heat treating above 1470°F followed by water quenching resulted in excellent ductility properties, whereas treating in the temperature range 930° to 1290°F resulted in loss of ductility with complete embrittlement occurring upon heat treating at 1020°F. In an effort to increase the aluminum content tolerable in Ti-Al binary alloys and to indicate continuity of properties with reference to composition and temperature range where ductility can be expected, alloys of 6 and 10Al were studied. An 8Al alloy prepared with iodide titanium was used to gain information on the effect of interstitial content on its embrittlement. Tensile testing, electron microscopy and electron or X-ray diffraction techniques were the principal tools employed.

Factors influencing ductility in the 7Al-3Mo alloy also were investigated.

Aging studies of the 5Al-2Ag and 5Al-7Ag alloys initiated in previous research work(5) were continued and some mechanical properties were determined.

II. EXPERIMENTAL PROCEDURE

A. Preparation of Alloys

1. Materials

With the exception of some Ti-8Al material which was prepared with iodide crystal bar, test ingots were made with sponge titanium of 120 BHN quality. The iodide titanium was vacuum annealed at pressures of less than 1 micron for 4 hours at 1560°F to remove hydrogen prior to the melting operation. The sponge was dried for 3 hours at 250°F to remove moisture which may have been picked up during storage.

The source and major impurities of the high purity elemental additions used are recorded in Table 1. The high melting additions, chromium, molybdenum, vanadium and zirconium, were charged in master alloys as particles -1/4 in. + 16 mesh in size. The compositions of these master alloys are given in Table 2.

2. Melting Technique

Both non-consumable and consumable electrode arc melting furnaces were employed. The non-consumable electrode furnace was used in instances where only small ingots were required. In such cases the ingots were inverted and remelted several times to insure homogeneity. The 6Al(2), iodide 8Al, 5Al-2Ag and 5Al-7Ag alloys were processed in this manner, the binary 6 and 8Al alloys as 200 gram buttons and the ternary alloys as 100 or 200 gram buttons. All other experimental alloys, following initial melting in the non-consumable furnace and forging to bar stock, were remelted in the consumable electrode furnace.

3. Forging Practice

An electric furnace was used for heating ingots to temperature. The 6Al(2) alloy was forged to 1/2 in. round at 1900°F. The 8Al alloy was forged to

TABLE 1

MATERIALS

Element	Form	Source	Purity	Impurities, %
Aluminum	Sheet	Steel Sales Co.	99.0%	
Chromium	Electrolytic Plate	Bureau of Mines	99.3%	0.6 O, 0.10 Fe, 0.02 S
Copper	Pellets			
Manganese	Electrolytic Plate	Bureau of Mines	99.9%	0.0035 Fe, 0.001 Cu, 0.018 S, 0.001 Pb, 0.0005 As
Molybdenum	Commercial Master Alloy	Climax Molybdenum Corp.		
Molybdenum	Chips	Fansteel Metallurgical Corp.	99.9%	0.005 Fe, 0.015 C, 0.045 O, 0.02 Co+Ni
Tin	Sheet	Vulcan Detinning Co.	99.99%	0.002 Fe, 0.0002 Sb
Vanadium	Commercial Master Alloy V-14Al	Electro Metallurgical Div. Union Carbide & Carbon Co.	98.2%	0.5 Fe, 0.4 Si, 0.03 C
Zirconium	Sponge	Bureau of Mines	99.5-99.8%	
Titanium	Iodide Bar	Foote Mineral Co.	99.9%	
Titanium	Sponge (BB)	E. I. Du Pont deNemours & Co.	120 BHN	0.07 Fe, 0.06 O, 0.04 C, 0.02 N, 15 ppm H

TABLE 2

MASTER ALLOYS

Composition	Comments
Ti-60Cr	Melting point below that of titanium; readily crushed.
Ti-31Mo	Commercially available. Added as chips. Improvement over unalloyed molybdenum. Double melting necessary to ensure homogeneity.
Mo-22Al	Melting point in vicinity of that of titanium; readily crushed.
V-14Al	Commercially available.
Zr-10Al	Melting point below that of titanium; readily crushed.

1/2 in. diameter bar stock at 2150°F. The Ti-Al-Ag alloys were forged to 3/8 in. round at 1750°F. The double melted 10Al alloy was forged initially to electrode dimensions at 2100°F and finish forged following the second melting operation at 2050°F.

All other experimental alloys were initially forged at 1900°F to 1 in. diameter electrodes. Following remelting, the alloys being investigated as forging materials were forged to 1/2 in. diameter rods in the range 1900° to 1800°F with the lower temperature being employed as the material reached final dimensions. The three alloys being studied for sheet applicability were forged in the 1900°-1800°F range to 1/2 in. plate and hot rolled to 0.075 in. sheet at 1750°F. Throughout this report forging temperature is defined as the temperature of the furnace.

4. Chemical Analysis

Results of chemical analyses on representative samples of each alloy are presented in Table 3. Hydrogen contents were determined by the hot extraction method, except where oxygen content was also determined. In this case the vacuum fusion method was used. In most instances the agreement between analyzed and nominal composition was quite good; as a consequence, each alloy is referred to in the text by its nominal composition.

B. Age Hardening Studies

Age hardening characteristics of the 5Al-2Ag and 5Al-7Ag alloys were evaluated in the temperature range 800° to 1020°F. Samples 3/8 in. in diameter by 1/4 in. sectioned from forged bar stock were used. All samples were given an initial 24 hour solution annealing treatment in evacuated Vycor bulbs at 1560°F. Aging heat treatments were also carried out in evacuated bulbs. Samples were water quenched following the solution anneal and the aging heat treatments.

C. Tensile Testing

Forging alloys were tested as standard 3 by 0.252 in. diameter specimens. The sheet alloys were tested as 0.060 in. sheet. Sheet specimens were 7 in. overall with 1/2 x 2 in. reduced section.

A Baldwin-Southwark 60,000 pound capacity tensile machine, equipped with an autographic stress-strain recorder, was used for the tensile testing phase of this investigation. Tests on the forging and sheet alloys were run at a loading rate of 1000 lbs/min, which amounts to an approximate strain rate of 0.001 in/in/min at room temperature over the interval in which the test piece is deforming elastically. Beyond the proportional limit the strain rate continually increases. Tests on the binary Ti-Al alloys were made at loading rates of 600 and 4000 lbs/min.

A Baldwin-Lima-Hamilton separable, microformer-type extensometer was employed to give plots of load versus strain, from which 0.2% offset yield strength and modulus of elasticity values were calculated. This extensometer can be used for both ambient and elevated temperature testing, and can be adapted for either bar or sheet specimens by a simple exchange of extension arms.

TABLE 3

CHEMICAL ANALYSIS OF EXPERIMENTAL ALLOYS

Nominal Composition	Analyzed Composition	Hydrogen Content, ppm
6Al (1)	5.98Al	233
6Al (2)	5.77Al	89
8Al (iodide)	8.23, 8.35Al	100
10Al	9.96Al	135
5Al-2Ag	5.49Al-2.11, 2.18Ag	
5Al-7Ag	5.32Al-6.24, 6.88Ag	
6Al-3Cu	6.14Al-2.91Cu	147
6Al-3Mo	6.14Al-3.08Mo	126
6Al-2Mo-1Mn	6.37Al-2.40Mo-1.13Mn	161
6Al-2Mo-1Cr	6.10Al-2.05Mo-0.97Cr	153
6Al-2Mo-1V	6.23Al-2.20Mo-0.90V	189
6Al-1Mo-1Mn-1V	6.54Al-1.10Mo-1.02Mn-0.90V	221
6Al-1Mo-1Mn-1Cr	6.16Al-1.00Mo-1.07Mn-1.00Cr	200
6Al-3Mo-4Sn	6.11Al-3.29Mo-3.56Sn	129
6Al-3Mo-4Zr	6.05Al-3.23Mo-3.91Zr	170
6Al-3Mo-2Sn-2Zr	5.82Al-3.35Mo-1.98Sn-2.22Zr	235
10Al-3Mo-4Zr	10.01Al-2.91Mo-4.16Zr	
6Al-2V-1Mo	6.30Al-2.10V-1.10Mo	73
6Al-2V-2Mo	6.19Al-1.98V-2.07Mo	71
6Al-3V-1Mo	6.37Al-3.00V-1.09Mo	106

For tests in the range -40° to 570°F the lower extension arm of the tensile assembly was equipped with a metal Dewar flask. Baths of various liquid media were used to maintain the test piece at the test temperature. The fluid media employed were the following:

-40°F - dry ice and acetone.
32°F - ice and water.
76° to 212°F - water.
212° to 480°F - mineral oil.

A 16 in. long by 2-1/2 in. I.D. Marshall furnace, regulated by a Wheelco controller, was used in the elevated temperature test work. Chromel-alumel thermocouples attached to the shoulders of the test piece indicated test temperature, which was held within ±5°F over the test section.

D. Creep and Creep-Rupture Testing

Eleven lever-arm creep test stands and five stress aging units were employe in the program. The forging alloy creep specimen was a modified version of the standard 0.252 in. diameter by 1.25 in. reduced section tensile specimen. The modification consisted of machining to much closer tolerances so that the effective reduced section was not more than about 4% of the 1.25 in. dimension. Sheet specimens were identical to those employed in tensile testing. Three recording chromel-alumel thermocouples were attached to each creep specimen by means of Inconel clips. Temperature was maintained within ±2°F over the test section. The extensometer used with the creep test stands for indication of extension of bar specimens was not readily adaptable for testing of sheet. Consequently, in the case of the sheet alloys, total extension was obtained on unbroken specimens by accurate optical measurements; extension with time data were not obtainable.

E. Stability Testing

Unbroken test specimens creep tested at 800° and 1020°F were tensile tested at room temperature to determine the influence of exposure at temperature under stress application. Specimens which maintain ductility values comparable to their unexposed counterparts are termed stable.

F. Impact Testing

Standard Charpy V-notch impact specimens were tested in the temperature range 32° to 212°F using a Riehle machine.

G. Bend Testing

Guided bend tests were made on samples of the three sheet modifications of the basic 6Al-4V composition. A Di-Acro unit was employed for this test work.

III. RESULTS AND DISCUSSION

A. Age Hardening of Ti-5Al-2Ag and Ti-5Al-7Ag Alloys

Following the 24 hour solution anneal at 1560°F, samples of each alloy were aged at 800°, 840°, 930° and 1020°F for times of 15 and 30 minutes, 1, 5, 24 and 48 hours. The aging results are recorded in Table 4 and diagrammatically presented as Figure 1. Over the temperature range investigated the alloys exhibited fairly typical aging response. At 800°F the 5Al-2Ag alloy reached higher peak hardness than the 5Al-7Ag alloy. The 5Al-2Ag alloy--although not indicating an overaging tendency at 800°F for times up to 48 hours--overages at shorter times with rise in temperature. It appears from the curves at 800° and 840°F that the 5Al-7Ag alloy overages quite rapidly. Disregarding the small aging peaks for the 5Al-7Ag alloy at the lower temperatures, the resemblance between the two alloys is quite marked at 840°F and above.

Microstructures of the two materials were similar. Representative structures of the 5Al-7Ag alloy are shown as Figures 2 and 3, which indicate copious precipitate occurring within the grains and at the grain boundaries.

Tensile specimens of the 5Al-2Ag alloy were aged for 24 hours at 800°F following the solution anneal, based on indications that this was the optimum heat treatment for maximum hardening response. Tensile properties at room temperature were the following:

Specimen No.	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area, %	Elongation %
(1)	125,000	105,000	41.5	15.0
(2)	132,000	114,500	20.5	12.0

A 5Al-7Ag alloy tensile specimen in the solution annealed condition: 1560°F-2 hours-water quenched, had the following properties: UTS = 126,300 psi, YS = 100,000 psi, RA = 27%, El = 12%. Aging for 48 hours at 930°F following the solution annealing treatment gave the following: UTS = 136,600 psi, YS = 117,200 psi, RA = 20%, El = 13%.

Creep-rupture tests made on the 5Al-2Ag alloy (H.T.: 1560°F-24 hrs-WQ, 800°F-24 hrs-AC) and the 5Al-7Ag alloy (H.T.: 1560°F-2 hrs-WQ, 930°F-48 hrs-AC) at 800°F indicated similar 100 and 500 hour rupture strengths of 68,000 and 64,000 psi, respectively; creep strengths to produce rates of 0.1%/hour and 0.01%/hour were 72,000 and 63,000 psi, respectively; see Figure 4. These alloys, based on the above values, are decidedly inferior to the binary 6Al alloy.

B. Results for Forging Alloys

1. Tensile Properties

Tensile characteristics of the alloys under investigation as forging materials were determined at room temperature, 620°, 800° and 1020°F. Duplicate

TABLE 4

AGING CHARACTERISTICS OF Ti-Al-Ag ALLOYS

Alloy*	Aging Temp. °F	Diamond Pyramid Hardness (20 Kg. Load) 0	15'	30'	1 hr.	5 hrs.	24 hrs.	48 hrs.
Ti-5Al-2Ag	800	275	280	282	290	292	330	328
	840	275	277	286	283	290	323	274
	930	275	274	282	284	304	289	279
	1020	275	302	287	306	308	271	282
Ti-5Al-7Ag	800	295	325	322	305	298	314	298
	840	295	315	291	292	295	306	297
	930	295	301	304	295	283	309	283
	1020	295	320	308	302	334	309	311

* Solution treated: 1560°F-24 hrs-WQ.

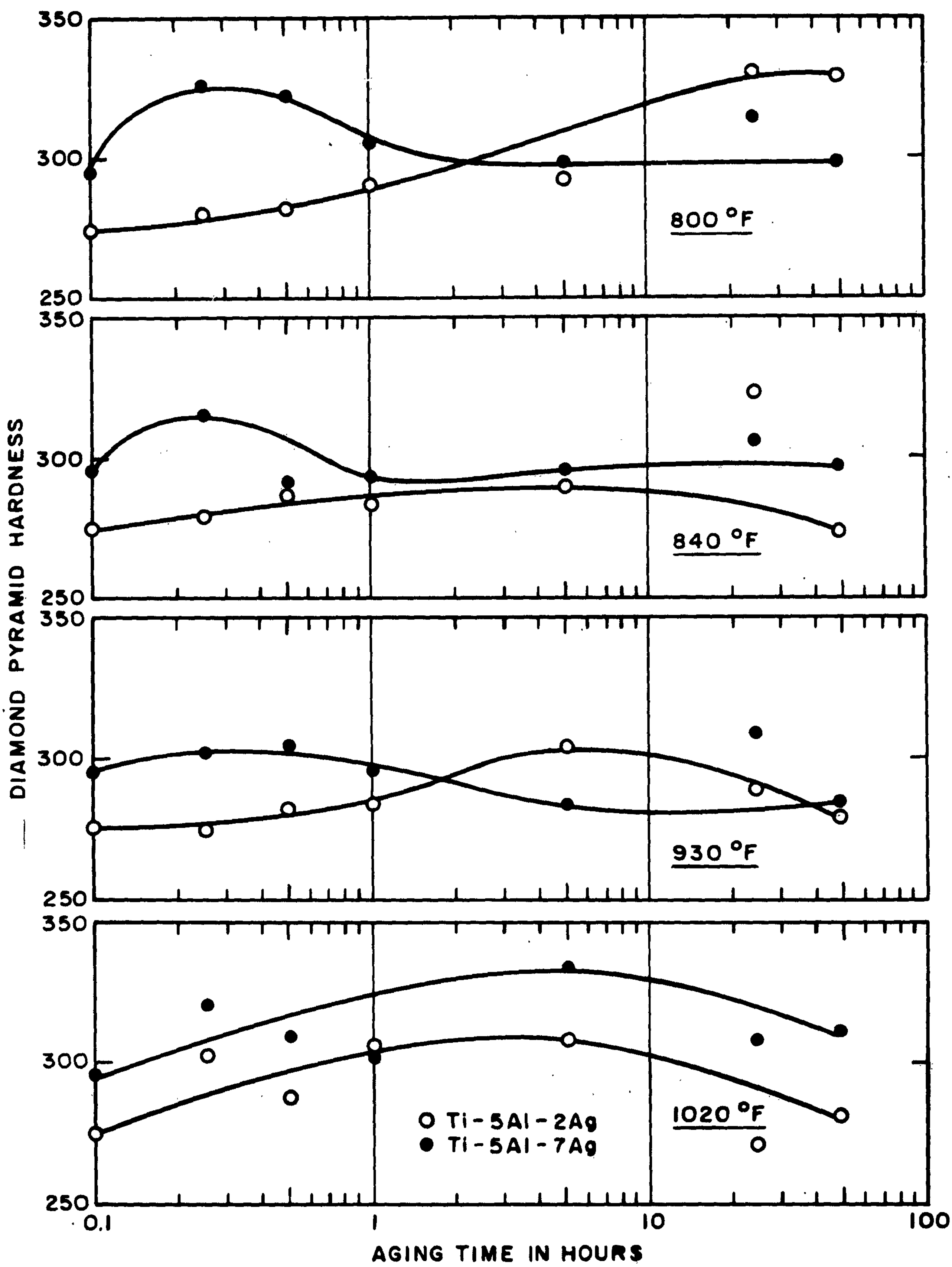
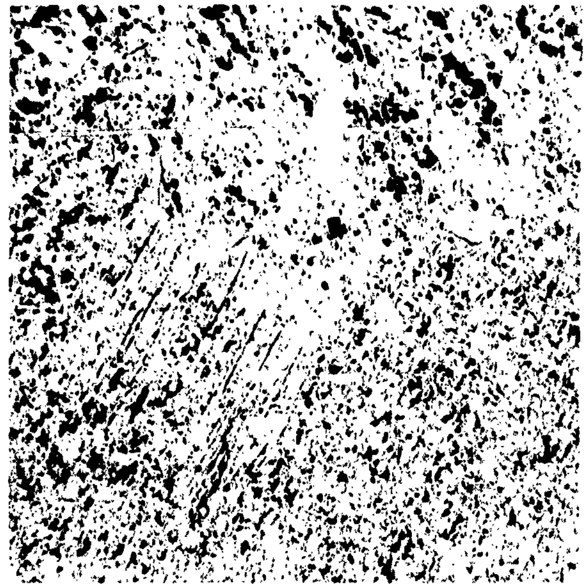


FIG. 1- AGE HARDENING CHARACTERISTICS OF Ti-5%Al-2%Ag AND Ti-5%Al-7%Ag ALLOYS



Neg. No. 11408 X 250

Fig. 2

Ti-5Al-7Ag: 1560°F-24 hrs-WQ, 930°F-48 hrs-WQ.

Neg. No. 11409 X 250

Fig. 3

Ti-5Al-7Ag: 1560°F-24 hrs-WQ, 930°F-500 hrs-WQ.

Etchant: 20% HF, 20% HNO_3, balance glycerine followed by 2% HF in water.

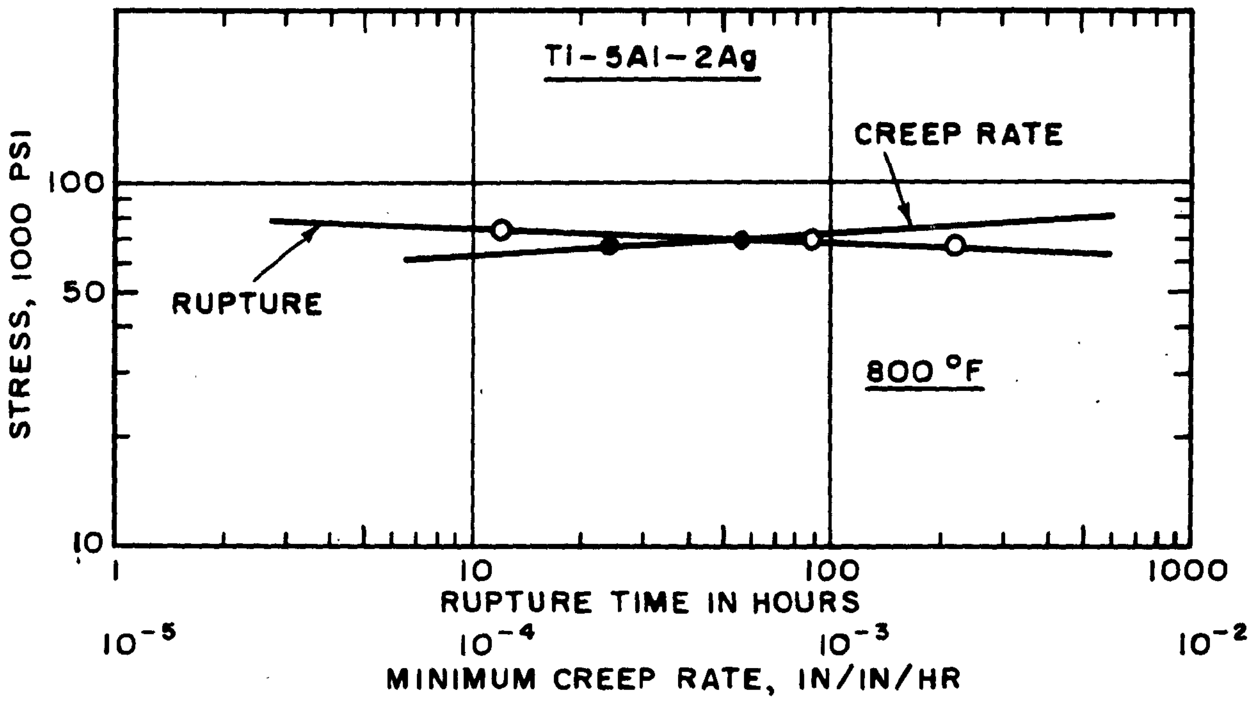


FIG. 4 - STRESS-RUPTURE AND MINIMUM CREEP RATE CURVES FOR Ti-5%Al-2%Ag ALLOY AT 800 °F

tests were made at each temperature on specimens given the following heat treatment: 1470°F-6 hours-air cool, 1020°F-24 hours-air cool. Test results are presented in Tables 5 to 9 and plotted as tensile property-temperature curves in Figures 5 to 15.

a. Room Temperature Properties

The alloy base material, Ti-6Al-3Mo, showed moderate room temperature strength (UTS = 145,000 psi, YS - 127,500 psi) coupled with acceptable ductility (RA = 28.5%, El = 12.5%). The structure of this alloy, as shown in Figure 16, was typical Widmanstätten α-β configuration. This structure was duplicated in most of the other alloys.

Replacement of 1% of molybdenum with manganese produced the Ti-6Al-2Mo-1Mn modification, which showed a 20,000 psi increase in ultimate strength with a 30,000 psi increase in yield strength. Strengthening was obtained at the expense of ductility, however, which dropped to values of 8.5% RA and 3.5% El. Hydrogen, present to the extent of 161 ppm, could possibly account for this poor showing, but other of the alloys in this group with higher hydrogen contents showed no ductility impairment. To check the influence of hydrogen, additional specimens were vacuum annealed to reduce hydrogen content to about 20 ppm. Results of tensile tests of vacuum annealed specimens are included as Table 6. The 6Al-2Mo-1Mn alloy was highly ductile as vacuum annealed; the other alloys were not significantly affected by the treatment. The structure of the 6Al-2Mo-1Mn alloy shown in Figure 17 was significantly different than that of the 6Al-3Mo alloy, consisting of patches of Widmanstätten α-β surrounded by equiaxed grains of α.

Substitution of 1% of either chromium or vanadium for molybdenum offered no improvement in ultimate strength, although vanadium raised the yield strength to 135,000 psi. The ductility of the chromium-bearing alloy was similar to the base material, whereas the vanadium-bearing alloy showed improvement (RA = 40%, El = 16%). Both alloys were structurally similar to the 6Al-3Mo alloy.

Further β complexing to give the alloys Ti-6Al-1Mo-1Mn-1V and Ti-6Al-1Mo-1Cr resulted in an increase in ultimate strength to a level in the vicinity of 155,000 psi. The former alloy had a much higher yield strength than the base alloy (147,000 psi compared to 127,500 psi), but also had markedly better ductility (RA = 44%, El = 17.5%). The ductility values of the latter alloy were slightly superior to those of the ternary base material. Figure 18 shows the structure of the vanadium-bearing alloy which was predominantly Widmanstätten α-β with islands of α. The structure of the 6Al-1Mo-1Mn-1Cr alloy was similar to that of the 6Al-3Mo alloy, but coarser.

Of the α complexed alloys, 6Al-3Mo-4Zr was only approximately 5000 psi stronger than the 6Al-3Mo base on the basis of both ultimate and yield strengths. Its ductility was nearly identical to the base composition. The 6Al-3Mo-4Sn and 6Al-3Mo-2Sn-2Zr alloys indicated an increase in ultimate strength to 155,000 psi. The yield strength of the 4Sn bearing alloy at 133,000 psi was slightly higher than that of the 6Al-3Mo alloy. The yield strength of the alloy with additions of 2Sn and 2Zr was still higher at 138,000 psi. Both of these alloys were comparable in ductilities to the alloy base. All three alloys were similar structurally to the 6Al-3Mo alloy.

TABLE 5

ROOM TEMPERATURE TENSILE TEST DATA FOR EXPERIMENTAL FORGING ALLOYS

Alloy*	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Modulus of Elasticity (psi x 10^{-6})	Fracture Stress psi	DPH
6Al-3Mo	144,000	127,000	28.5	12.5	18.3	180,500	333
	146,000	128,000	28.5	12.5	17.3	181,500	337
6Al-2Mo-1Mn	166,000	158,500	5.5	3.0[a]	18.6	175,500	367
	167,400	157,600	8.5	3.5	19.0	183,500	356
6Al-2Mo-1Cr	147,500	130,200	21.0	11.0	17.3	171,000	333
	147,500	129,000	38.5	16.5	17.9	197,500	333
6Al-2Mo-1V	140,000	136,500	--	a, b	18.4		333
	149,000	133,300	40.0	16.0	18.4	199,500	321
6Al-1Mo-1Mn-1V	156,500	147,000	44.0	17.5	15.0	220,000	357
6Al-1Mo-1Mn-1Cr	150,300	135,400	29.5	14.0	16.7	188,500	345
	156,800	141,000	33.5	13.5	15.6	204,000	339
6Al-3Mo-4Sn	157,800	133,200	21.0	13.5	14.8	193,200	343
	154,000	132,000	25.0	15.5	15.2	194,500	341
6Al-3Mo-4Zr	150,800	132,000	28.0	12.0	16.2	208,000	327
10Al-3Mo-4Zr	80,500[c]	---	--	--	22.0	80,500	381
	104,000[c]	---	--	--	19.6	104,000	386
6Al-3Mo-2Sn-2Zr	154,500	139,200	--	a, b	18.2	---	319
	157,400	137,300	26.5	13.0	15.4	198,500	339
6Al-3Cu	136,000[c]	---	--	--	19.0	136,000	367
	147,000	135,000	5.5	5.0	16.5	156,000	356

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.
a Forging seam.
b Failure occurred outside gage marks.
c Failure occurred in threaded section.

TABLE 6

ROOM TEMPERATURE TENSILE DATA FOR VACUUM ANNEALED EXPERIMENTAL FORGING ALLOYS

Alloy	Heat Treatment	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Modulus of Elasticity (psi x 10-6)	Fracture Stress psi
6Al-3Mo	1	134,800	---	9.0	7.0[a]	--	---
		143,000	---	16.0	12.0	--	163,800
	2	144,000	123,000	26.5	13.0	18.7	179,000
		142,000	125,300	17.0	8.0[a]	16.3	165,500
6Al-2Mo-1Mn	1	156,200	135,000	23.0	13.0	21.0	197,000
	2	149,000	139,000	23.0	12.0	18.1	---
6Al-2Mo-1Cr	1	148,400	---	27.5	15.0	--	182,500
		150,000	---	15.5	12.0	--	173,000
	2	150,200	131,300	32.5	15.0	15.3	199,000
		150,200	126,300	25.0	12.0	13.0	186,000
6Al-2Mo-1V	1	137,300	---	34.0	13.0	--	179,000
	2	144,600	125,200	35.0	13.0	17.9	---
		142,500	121,000	33.0	13.0	17.7	---
6Al-1Mo-1Mn-1Cr	1	152,200	134,800	19.5	12.5	16.5	185,000
		153,500	---	29.0	17.0	--	191,500
	2	145,300	137,200	30.0	12.0	22.0	200,000
		145,300	138,000	25.0	11.0	21.0	187,500
6Al-1Mo-1Mn-1V	1	158,000	---	15.5	9.0	--	157,900
	2	160,500	148,000	32.5	15.0	17.0	209,000
		148,000	146,000	5.5	3.0[a]	16.1	156,500

TABLE 6 (continued)

ROOM TEMPERATURE TENSILE DATA FOR VACUUM ANNEALED EXPERIMENTAL FORGING ALLOYS

Alloy	Heat Treatment	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Modulus of Elasticity (psi x 10^{-6})	Fracture Stress psi
6Al-3Mo-4Sn	1	155,500	---	23.0	14.0	--	193,500
	2	163,200	140,000	22.0	13.0	18.1	199,000
		151,700	140,000	4.0	3.0[a]	18.2	161,400
6Al-3Mo-2Sn-2Zr	1	159,000	---	22.0	11.0	--	185,000
	2	164,300	146,300	25.0	14.0	15.0	200,000
6Al-3Cu	1	147,000	---	2.5	2.0	--	150,800
		149,300	---	c	c	--	---
	2	151,000	132,400	23.0	11.0	23.5	---
		138,500	136,000[b]	c	c	18.0	---

1 VA 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

2 VA 1560°F-4 hrs-AC, 1020°F-24 hrs-AC.

a Forging seam.

b 0.1% offset yield strength.

c Failed outside gage marks.

TABLE 7

TENSILE TEST DATA FOR EXPERIMENTAL FORGING ALLOYS AT 620°F

Alloy*	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Modulus of Elasticity (psi x 10-6)	Fracture Stress psi
6Al-3Mo	109,000	89,000	49.5	16.0[a]	16.5	139,500
	112,000	95,600	66.5	22.5	13.8	197,000
6Al-2Mo-1Mn	123,000	92,000	47.5	16.0	19.5	183,500
	121,500	99,000	21.0	9.0[a]	16.1	125,000
6Al-2Mo-1Cr	111,000	88,400	49.0	17.5	12.7	162,800
	109,200	85,300	59.5	20.0	12.2	188,400
6Al-2Mo-1V	108,500	89,000	53.5	17.0	14.0	172,000
	104,000	78,000	56.5	20.0	16.7	182,500
6Al-1Mo-1Mn-1V	116,200	91,000	48.5	18.0	15.8	175,000
	120,000	95,800	54.0	18.5	13.0	196,500
6Al-1Mo-1Mn-1Cr	112,000	87,500	44.0	17.0	10.7	167,200
	113,000	93,500	46.5	17.5	--	173,000
6Al-3Mo-4Sn	121,000	93,000	43.5	20.0	--	195,800
	122,200	94,000	43.5	20.0	14.1	179,600
6Al-3Mo-4Zr	112,800	---	37.0	17.0	--	156,000
10Al-3Mo-4Zr	160,000	134,200	25.0	14.0	13.5	209,000
	150,500	127,000	15.0	11.0[a]	16.3	176,500
6Al-3Mo-2Sn-2Zr	122,000	96,700	36.0	15.5	--	164,000
	121,200	95,300	50.5	17.0	13.0	187,000
6Al-3Cu	110,500	95,000	32.5	11.0	10.3	157,200
	117,000	100,000	32.5	13.0	14.3	161,000

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

a Fracture ...

TABLE 8

TENSILE TEST DATA FOR EXPERIMENTAL FORGING ALLOYS AT 800°F

Alloy*	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Modulus of Elasticity (psi x 10^{-6})	Fracture Stress psi
6Al-3Mo	110,000	90,400	63.5	19.0	11.0	182,500
	94,500	78,800	60.5	19.0	9.0	168,500
6Al-2Mo-1Mn	118,000	102,000	56.5	16.5	16.3	187,500
	116,500	96,500	51.5	16.5	17.4	178,500
6Al-2Mo-1Cr	102,300	80,700	59.0	18.5	--	169,000
	98,800	75,500	58.5	18.5	--	162,600
6Al-2Mo-1V	99,500	83,500	59.5	15.0	--	174,000
	98,500	79,800	62.0	17.0	--	169,000
6Al-1Mo-1Mn-1V	106,500	88,200	44.5	15.0[a]	--	164,500
	110,500	93,500	46.0	14.5	--	154,500
6Al-1Mo-1Mn-1Cr	106,000	84,800	48.5	17.5	9.5	157,400
	103,300	80,500	50.5	12.0	--	160,500
6Al-3Mo-4Sn	122,000	94,400	38.0	17.0	13.1	167,200
	104,000	91,000	42.0	14.5	11.5	164,000
6Al-3Mo-4Zr	102,500	92,800	49.0	13.0	12.0	183,500
10Al-3Mo-4Zr	151,300	131,200	25.0	13.5	--	193,000
	145,800	121,500	15.7	11.0[a]	11.5	167,200
6Al-3Mo-2Sn-2Zr	115,600	94,300	52.0	17.0	12.0	189,500
	112,000	87,300	49.5	16.0	12.0	177,000
6Al-3Cu	105,800	93,000	37.0	14.5	--	154,500
	126,000	113,500	33.0	13.0	9.5	173,000

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

a Forging seam.

TABLE 9

TENSILE TEST DATA FOR EXPERIMENTAL FORGING ALLOYS AT 1020°F

Alloy*	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Modulus of Elasticity (psi x 10^{-6})	Fracture Stress psi
6Al-3Mo	78,800	61,500	39.5	14.0[a]	8.0	---
	86,600	69,300	60.5	21.0	10.0	111,000
6Al-2Mo-1Mn	87,400	70,500	59.5	20.0	--	119,000
	93,000	70,200	51.5	21.0	8.0	106,000
6Al-2Mo-1Cr	88,200	68,200	65.5	22.0	10.5	121,000
	86,000	69,700	60.5	21.5	8.0	113,200
6Al-2Mo-1V	82,500	66,000	53.0	17.0	--	120,500
	87,500	70,500	57.5	19.0	--	124,500
6Al-1Mo-1Mn-1V	93,500	73,000	58.5	18.0	9.0	137,000
	89,500	69,600	56.5	22.0	9.0	124,000
6Al-1Mo-1Mn-1Cr	88,800	65,800	58.0	21.0	9.0	107,000
	89,200	65,700	51.0	17.5	10.5	101,000
6Al-3Mo-4Sn	107,200	86,000	46.0	18.0	--	141,000
	105,000	84,800	49.0	19.0	--	139,000
6Al-3Mo-4Zr	90,000	72,000	53.5	13.0	11.0	125,000
10Al-3Mo-4Zr	149,000	124,800	39.5	17.5	10.5	201,000
	136,200	115,400	38.5	19.0	--	180,500
6Al-3Mo-2Sn-2Zr	107,600	85,700	56.0	18.5	10.5	155,200
	104,000	81,300	57.0	21.0	--	142,600
6Al-3Cu	104,000	81,100	47.5	20.0	--	131,000
	114,500	92,000	33.5	15.5	9.5	135,000

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

a Forging seam.

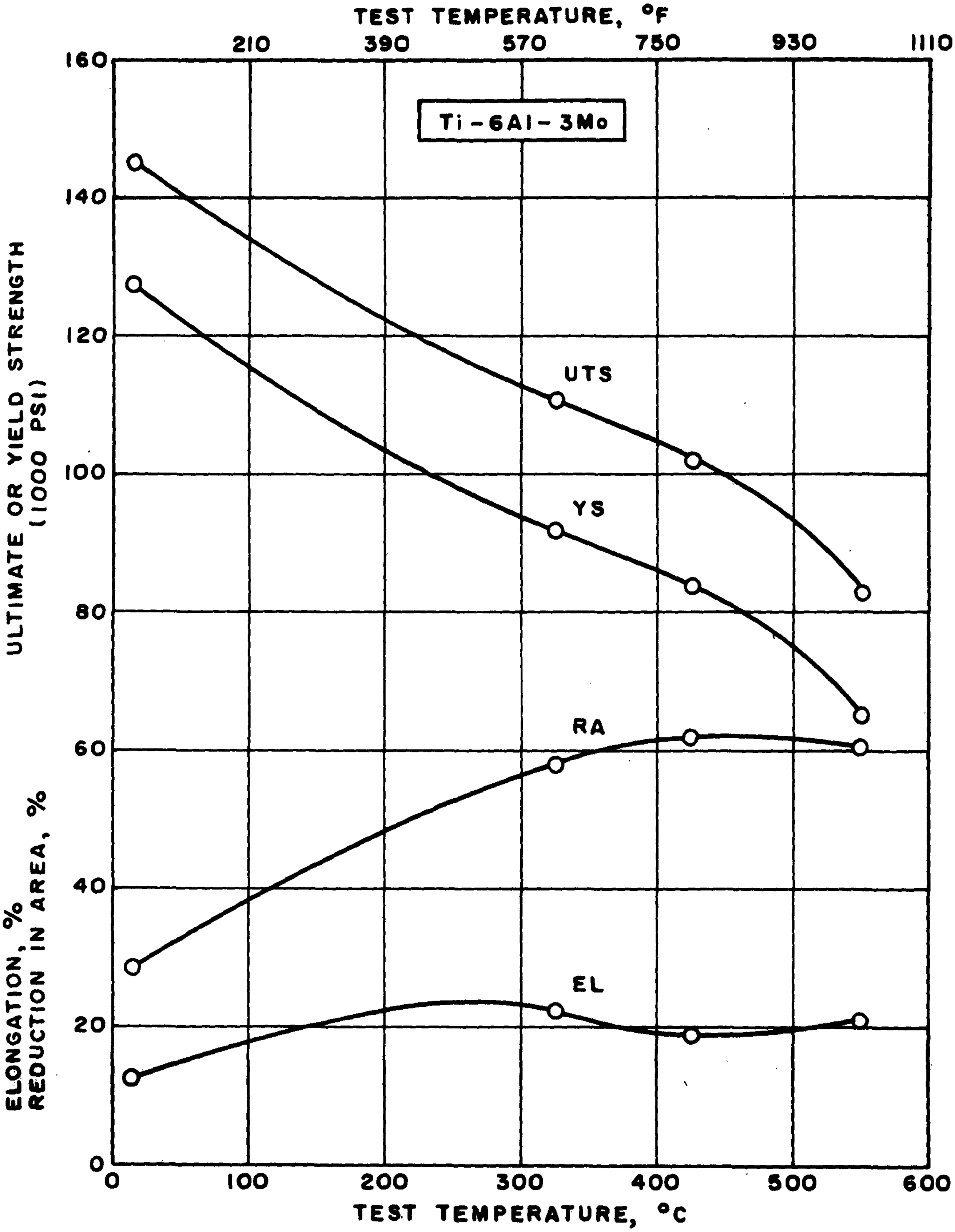


FIG. 5 - TENSILE PROPERTIES OF Ti-6%Al-3%Mo ALLOY

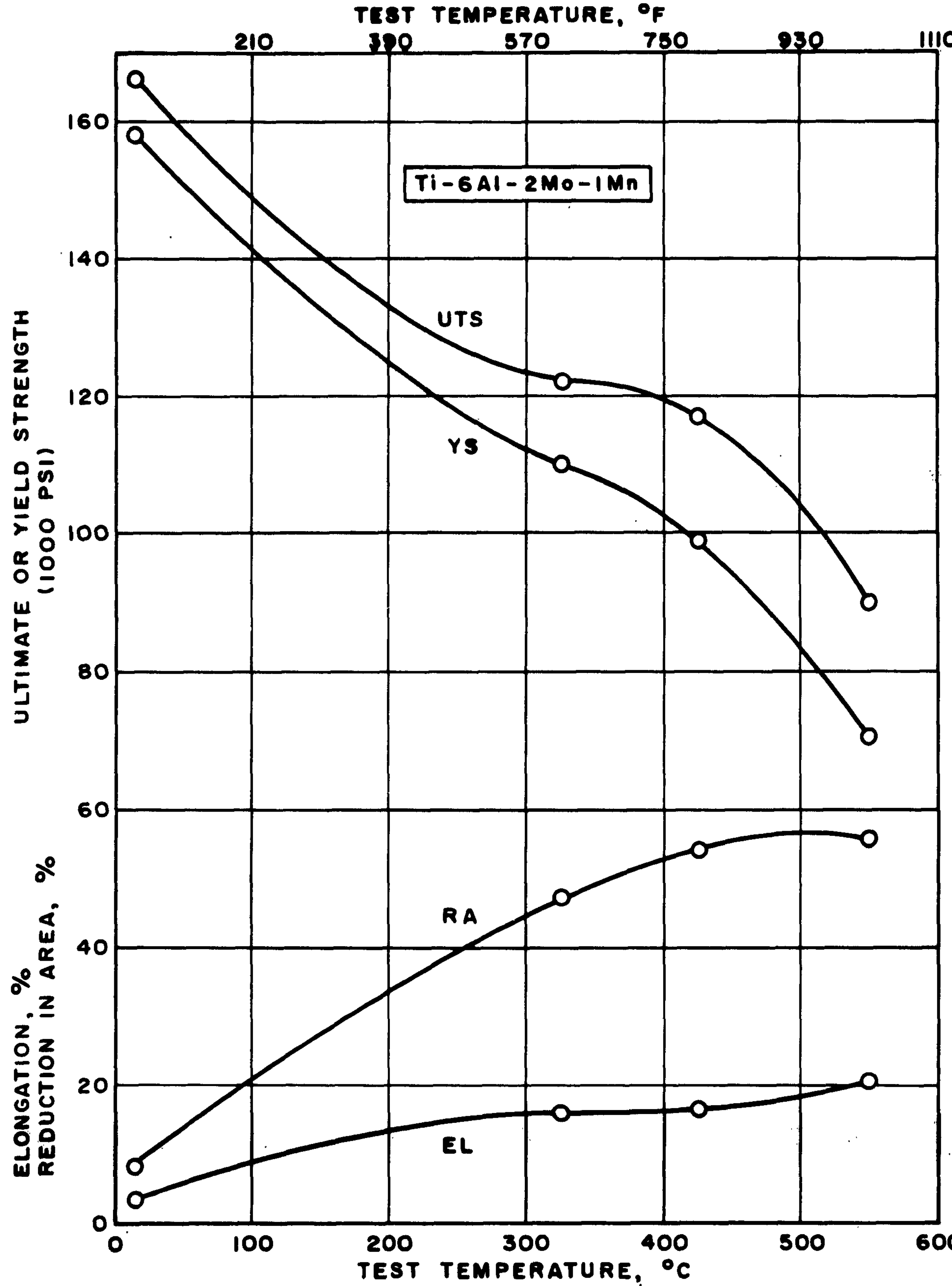


FIG. 6 - TENSILE PROPERTIES OF Ti-6%Al-2%Mo-1%Mn ALLOY

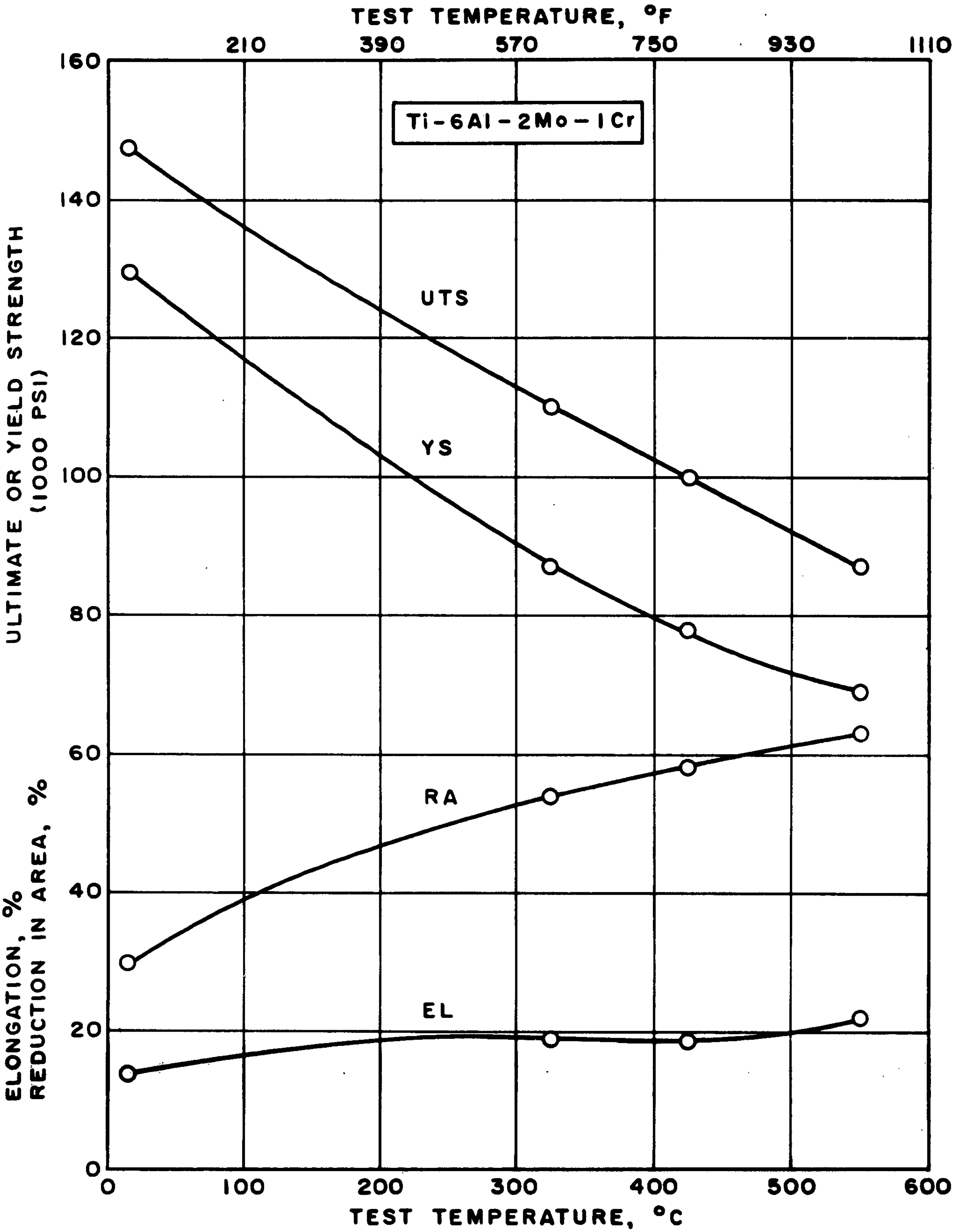


FIG. 7 - TENSILE PROPERTIES OF Ti-6%Al-2%Mo-1%Cr ALLOY

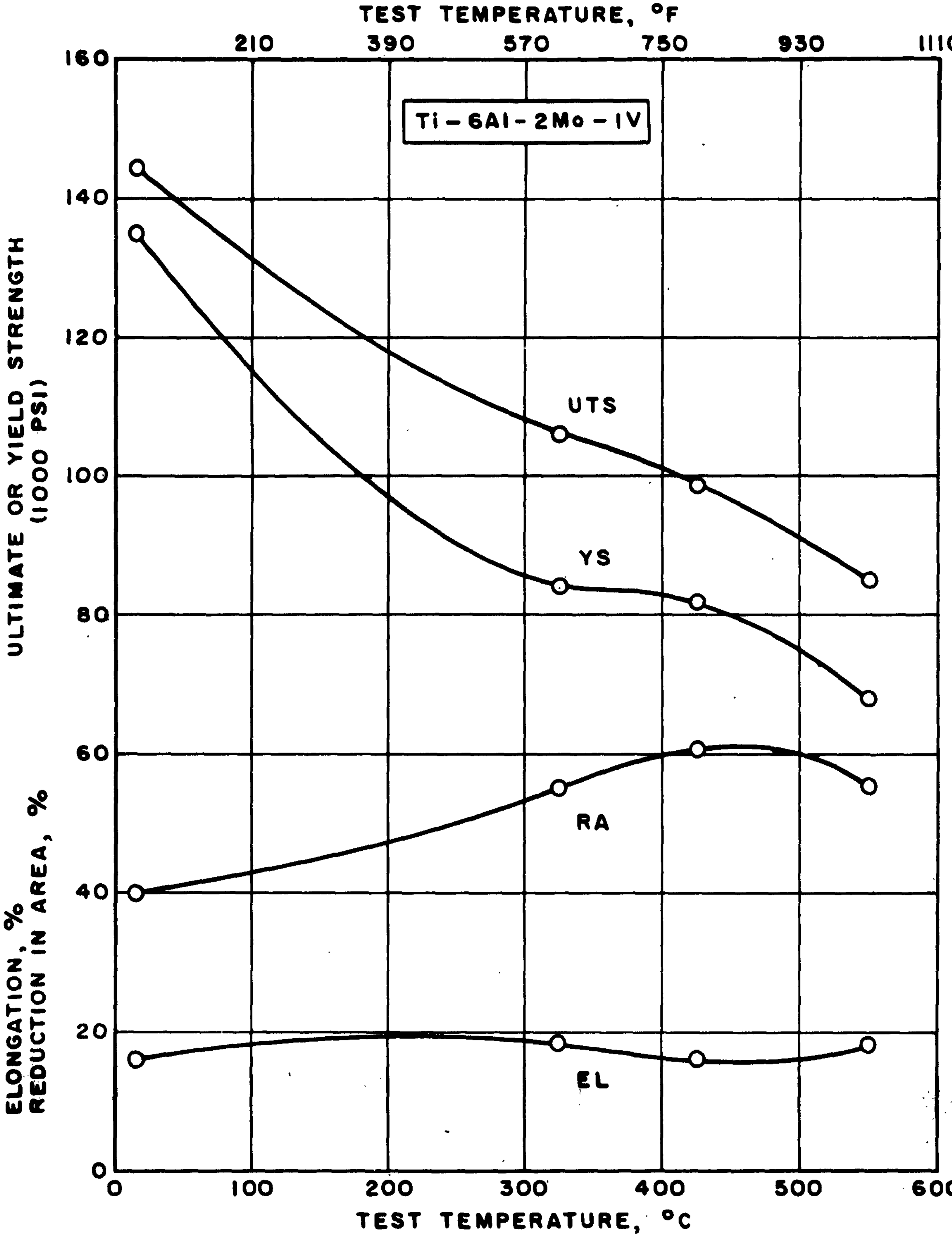


FIG. 8 - TENSILE PROPERTIES OF Ti-6%Al-2%Mo-1%V ALLOY

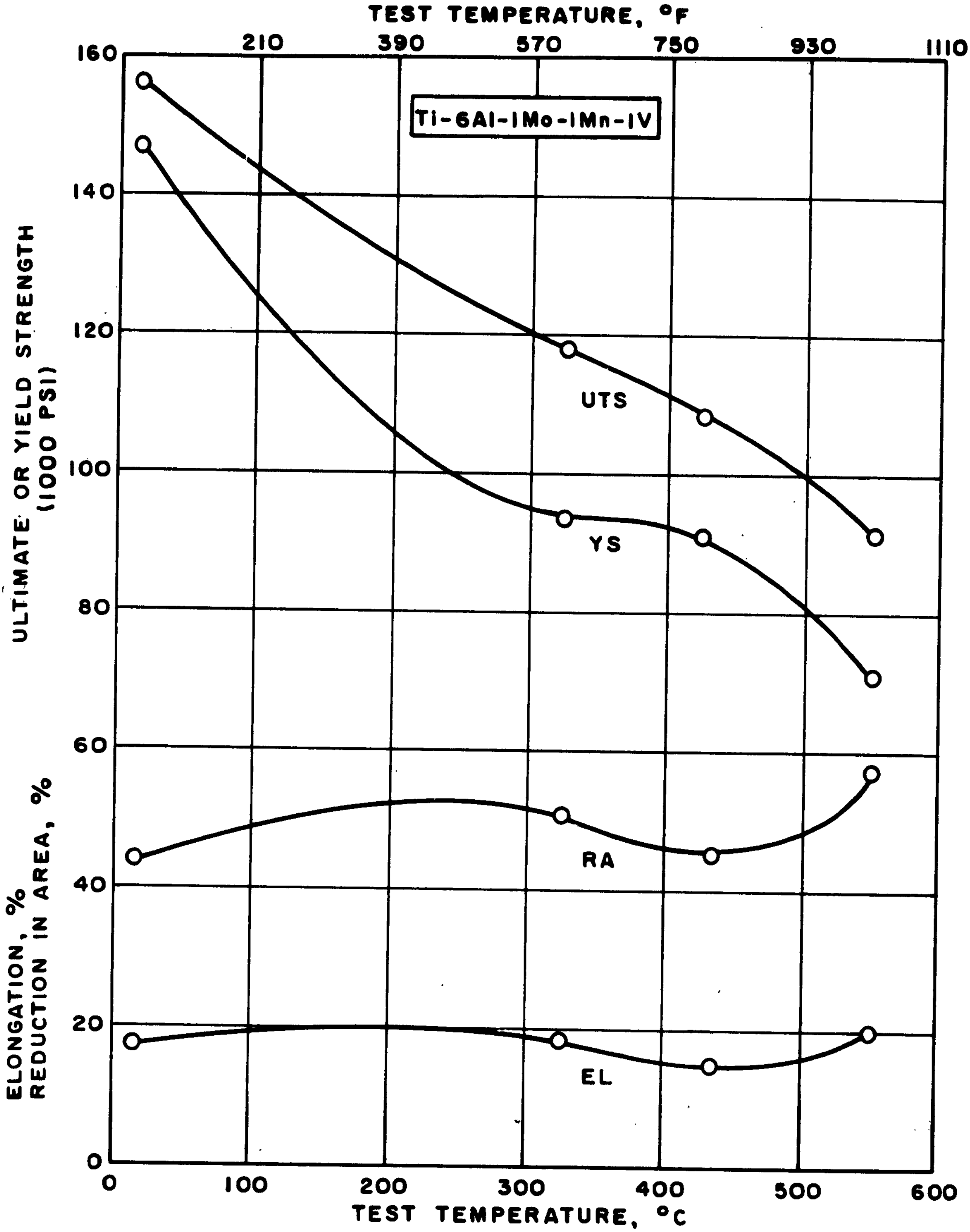


FIG. 9 – TENSILE PROPERTIES OF Ti-6%Al-1%Mo-1%Mn-1%V ALLOY

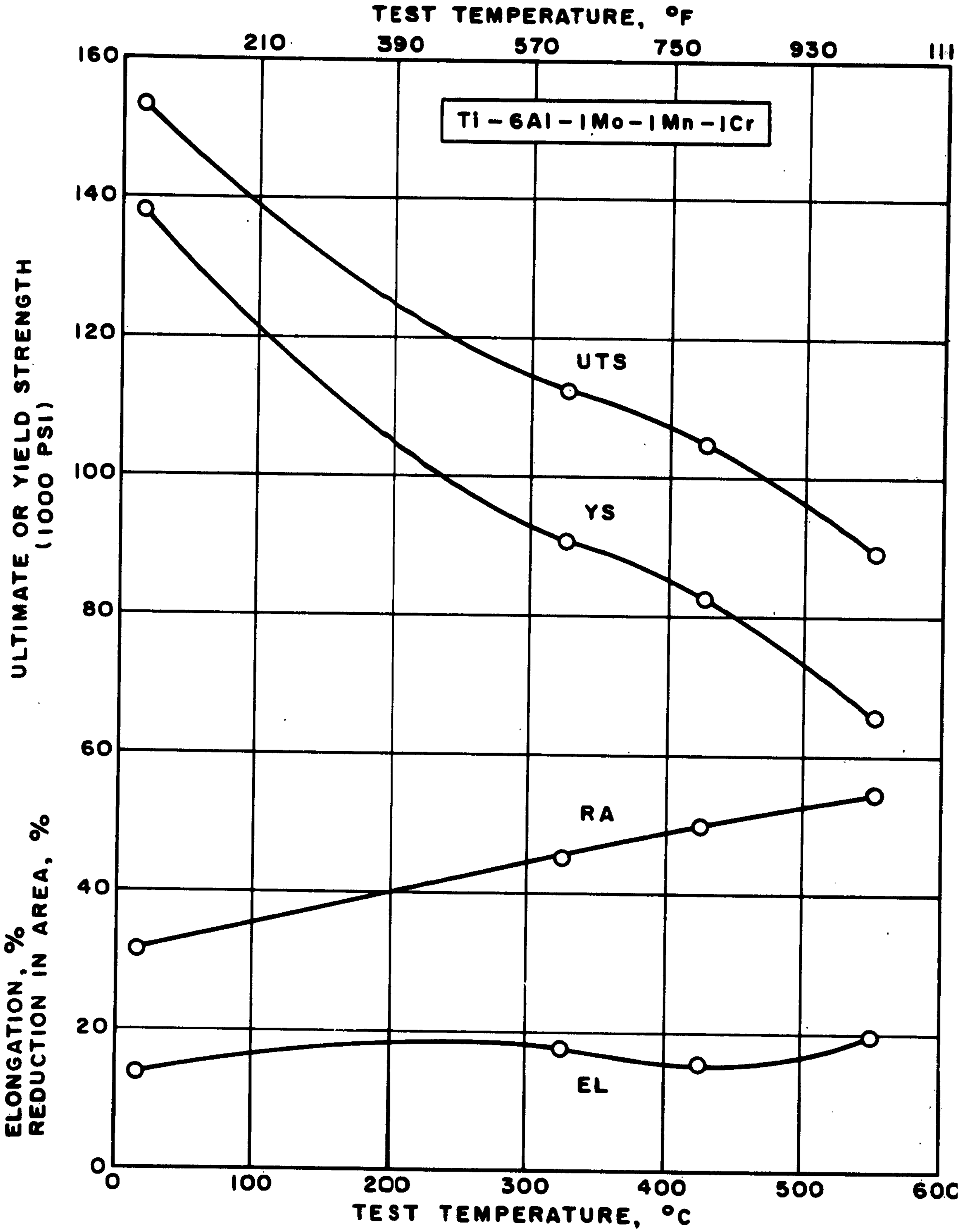


FIG. 10 - TENSILE PROPERTIES OF Ti-6%Al-1%Mo-1%Mn-1%Cr ALLOY

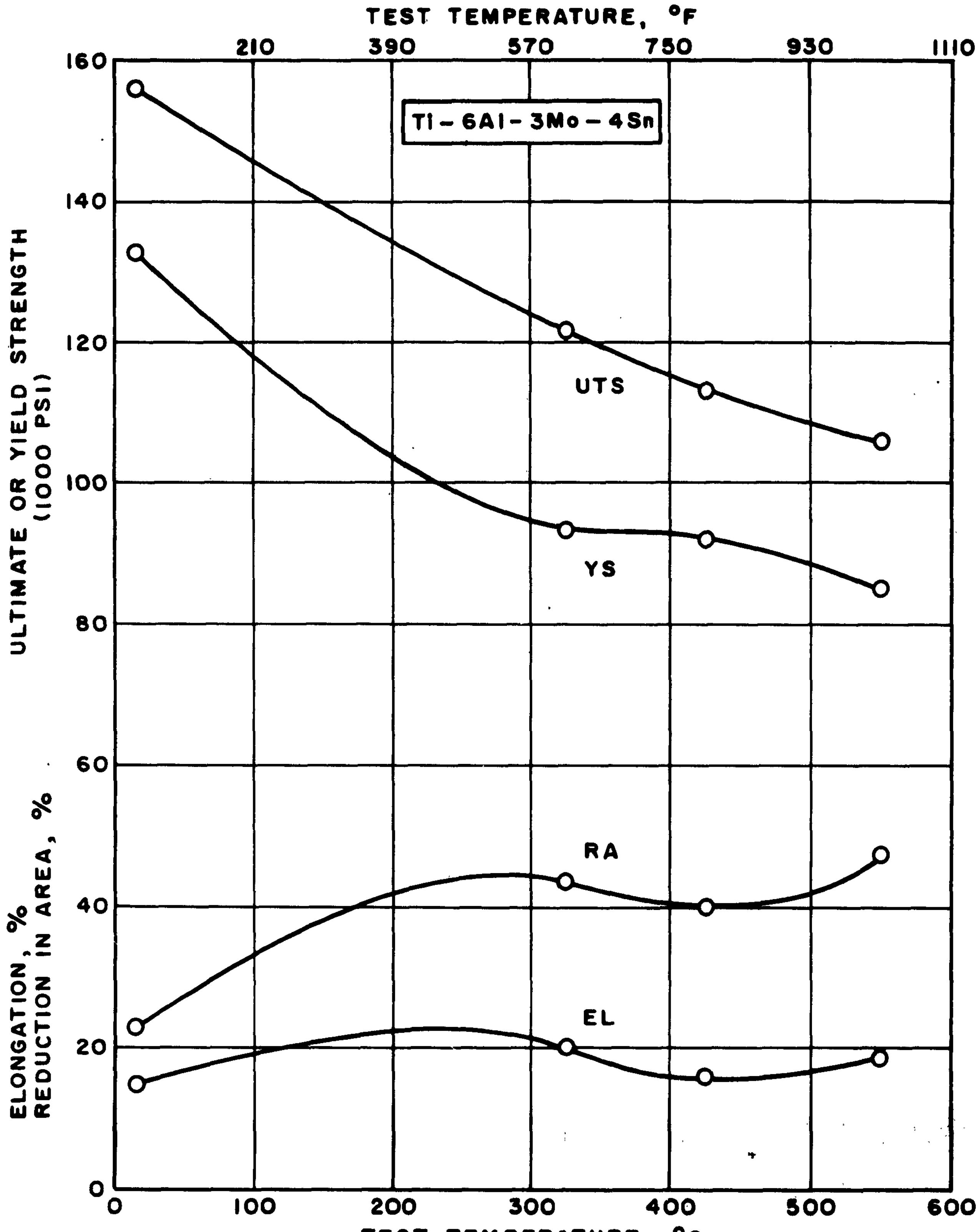


FIG. 11 – TENSILE PROPERTIES OF Ti-6%Al-3%Mo-4%Sn ALLOY

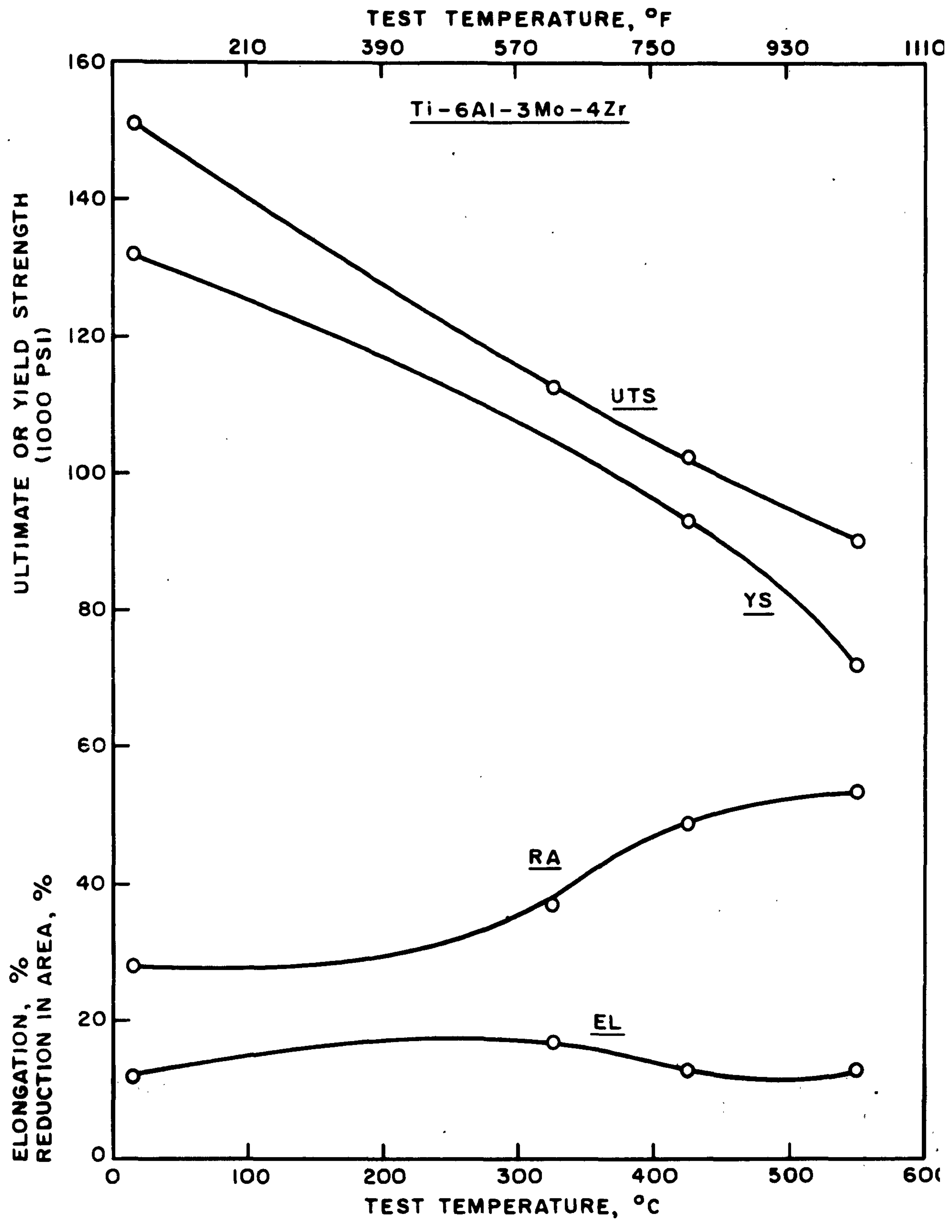


FIG. 12 – TENSILE PROPERTIES OF Ti-6%Al-3%Mo-4%Zr ALLOY

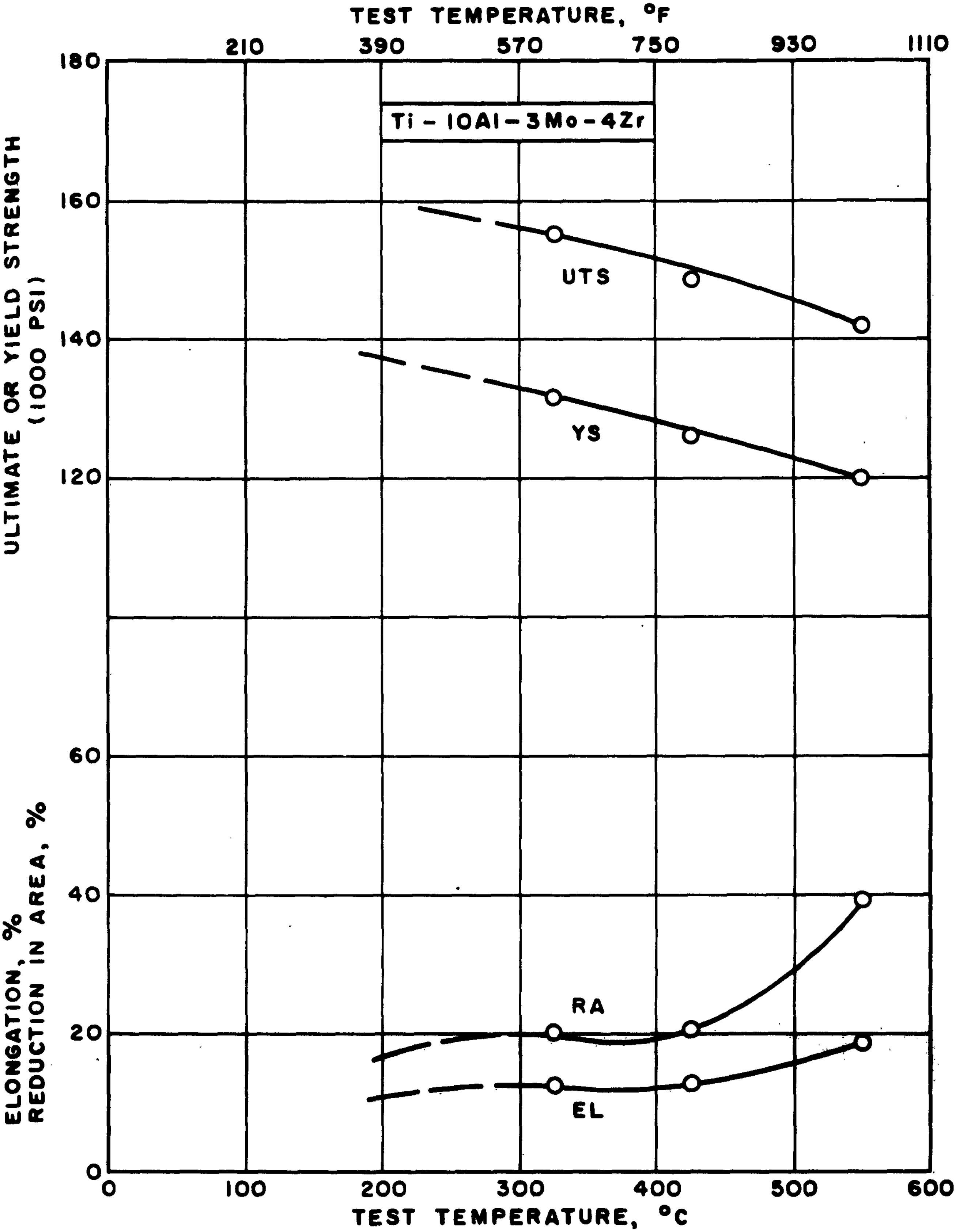


FIG. 13 – TENSILE PROPERTIES OF Ti-10%Al-3%Mo-4%Zr ALLOY

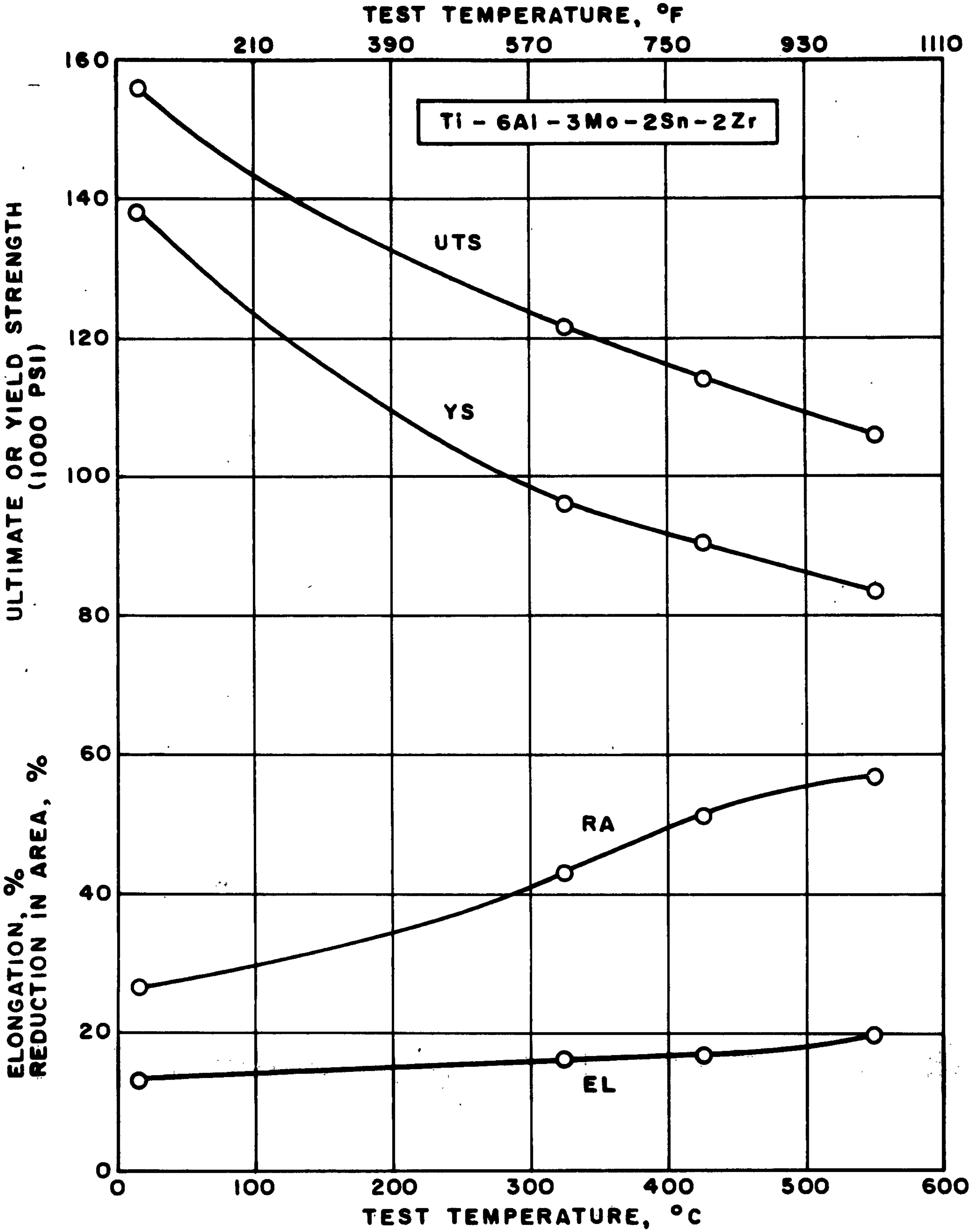


FIG. 14 – TENSILE PROPERTIES OF Ti-6%Al-3%Mo-2%Sn-2%Zr ALLOY

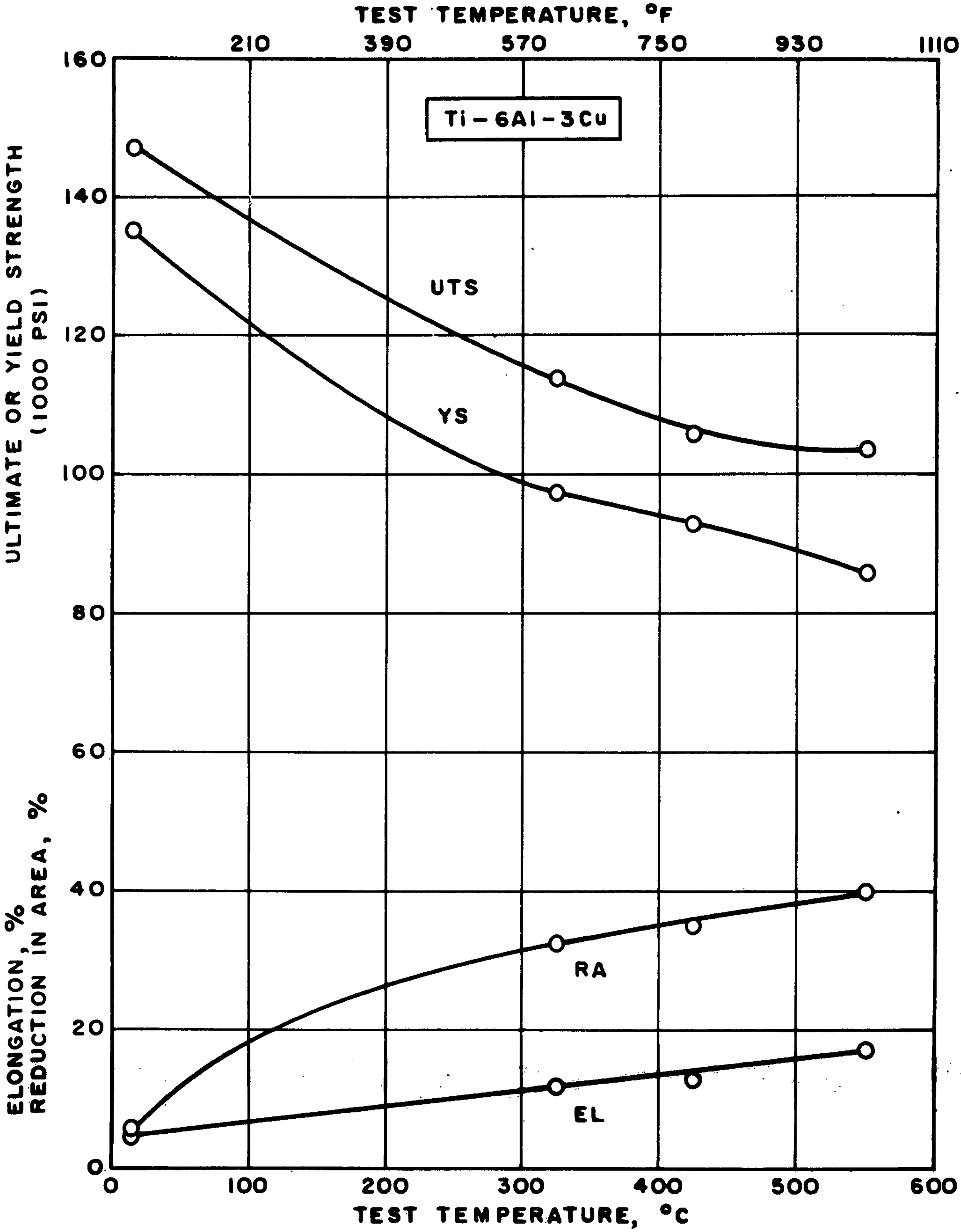
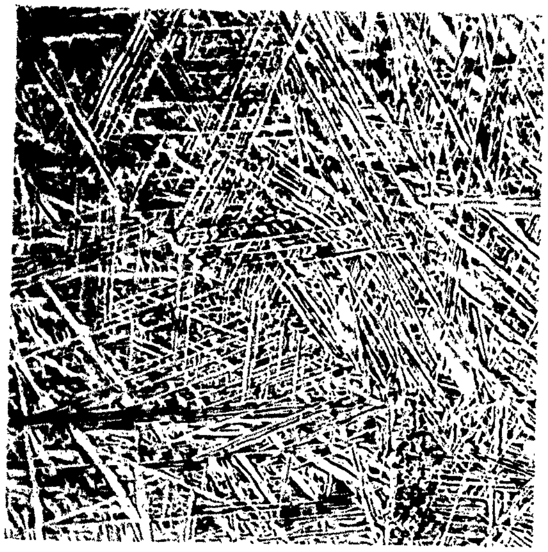


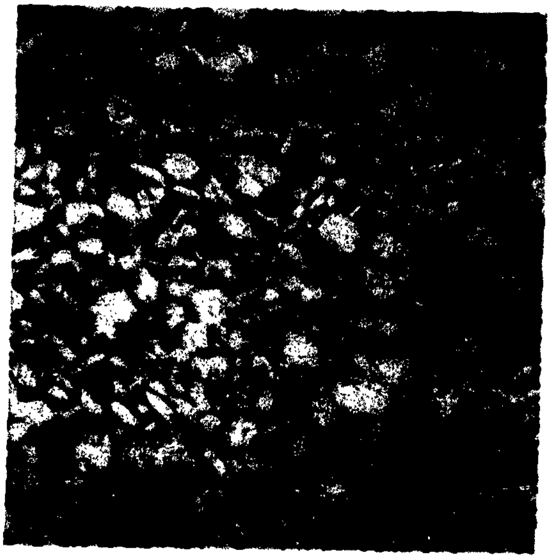
FIG. 15 - TENSILE PROPERTIES OF Ti-6%Al-3%Cu ALLOY



Neg. No. 11055 X 750

Fig. 16

Ti-6Al-3Mo, 1470°F-6 hrs-AC, 1020°F-24 hrs-AC. α + β.



Neg. No. 11046 X 750

Fig. 17

Ti-6Al-2Mo-1Mn, 1470°F-6 hrs-AC, 1020°F-24 hrs-AC. α + β.

Neg. No. 11049 X 750

Fig. 18

Ti-6Al-1Mo-1Mn-1V, 1470°F-6 hrs-AC, 1020°F-24 hrs-AC. α + β.

Etchant: 60 cc glycerine, 20 cc HNO_3, 20 cc HF.

A 10Al-3Mo-4Zr alloy, which was unintentionally prepared, was brittle at room temperature as indicated by failure in the threaded sections of the test pieces. Structurewise, this material was very coarse Widmanstätten.

The Ti-6Al-3Cu alloy, which is an α + compound type material, had room temperature strength properties very similar to those of the 6Al-3Mo alloy, but showed very poor ductility (one specimen failed in the threads, while the other showed an RA of 5.5% and an El of 5%). The $\alpha + Ti_2Cu$ structure for these specimens is shown in Figure 19. It was suspected that the low ductility of the 6Al-3Cu specimens was due to this Widmanstätten type structure. Much better ductility would be expected from an equiaxed structure. Therefore, the 6Al-3Cu specimens represented in Table 6 were finish forged at a lower temperature than that used to produce the structure shown in Figure 19a, i.e., 1750°F compared to 1800°F. However, for unknown reasons only one of these specimens had the desired structure. This was a ductile specimen and its structure is shown in Figure 19b. The other specimens had Widmanstätten type structures.

b. Elevated Temperature Properties

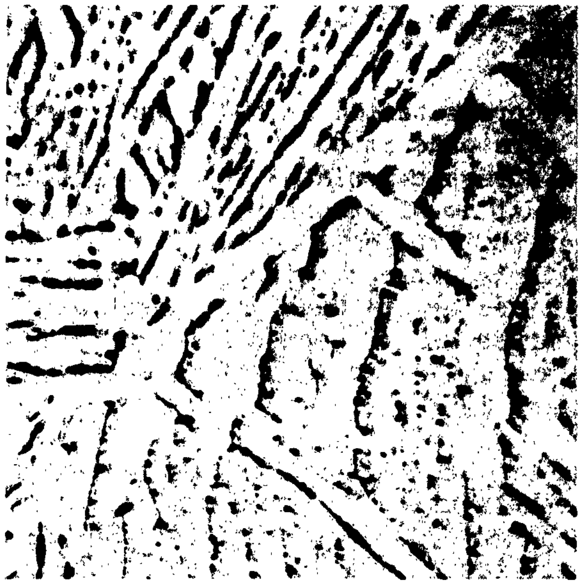
At 620°F and above all the alloys had acceptable ductility values. In general, the ductilities increased with temperature. The α complexed alloys were somewhat stiffer than the 6Al-3Mo and did not show as great a change with temperature as did the β complexed alloys.

The 10Al-3Mo-4Zr alloy was significantly superior on the basis of strength considerations at 620°F to the other materials, having an ultimate strength at this temperature of 155,000 psi and a yield strength of 130,000 psi. The superior performance of this alloy was maintained at 800° and 1020°F and is a reflection of the influence of aluminum in elevated temperature strengthening.

At 620°F on the 6Al level the 6Al-2Mo-1Mn β complexed alloy and two α complexed alloys, 6Al-3Mo-4Sn and 6Al-3Mo-2Sn-2Zr, were strongest at 122,000 psi. This stress level was approximately 10,000 psi higher than that found in the 6Al-3Mo alloy. With the exception of the 6Al-2Mo-1V alloy, which had an ultimate of 106,000 psi, the other alloys were close to the base at 111,000 psi.

At 800°F the 6Al-2Mo-1Mn alloy with an ultimate strength of 117,000 psi again showed the highest strength, followed by the α complexed 6Al-3Mo-4Sn and 6Al-3Mo-2Sn-2Zr alloys at 114,000 psi. The remaining α complexed alloy, 6Al-3Mo-4Zr, and the other β complexed alloys together with the base material were grouped between 100,000 and 106,000 psi. The 6Al-3Cu alloy was quite strong at this temperature with an ultimate strength of 116,000 psi.

The strength of the 6Al-3Mo alloy dropped precipitously at 1020°F to about 82,000 psi. The alloys with chromium, manganese or vanadium and the α complexed 6Al-3Mo-4Zr alloy were grouped around 90,000 psi. The α complexed 6Al-3Mo-4Sn and 6Al-3Mo-2Sn-2Zr alloys showed their elevated temperature potential with ultimate strengths at 106,000 psi. The copper-bearing alloy was strongest at this temperature with an ultimate of 109,000 psi.



Neg. No. 11053 X 750

a

1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

Neg. No. 11582 X 750

b

VA 1560°F-4 hrs-AC, 1020°F-24 hrs-AC.

Fig. 19 - Ti-6Al-3Cu. α + Ti_2Cu.

2. Creep-Rupture Properties

The forging alloys were creep tested at 800°F and creep-rupture tested at 1020°F. Three creep tests were made at 800°F for each alloy. In most instances one test was conducted at a stress calculated to cause failure within 300 hours. The remaining tests were run at lower stress levels and were terminated without failure at 300 hours. Creep-rupture tests at 1020°F were made to produce sufficient data points for reliable extrapolation of the stress vs. rupture time curves to 500 hour rupture lives and to enable preparation of stress vs. minimum creep rate curves. These curves are presented in Figures 20 to 29. Test data are contained in Tables 10 and 11. A summary of the rupture strength and creep rate data is also included as Table 12.

On the basis of stress-rupture strength at 1020°F weakening resulted from complexing the β phase by substitution of chromium, manganese or vanadium for part of the molybdenum of the 6Al-3Mo base. The 6Al-2Mo-1Cr alloy most closely approached the base material, having 100 and 500 hour rupture strengths of 32,000 and 22,000 psi, respectively, as compared to 40,000 and 33,000 psi for the 6Al-3Mo alloy. The α complexed alloys, 6Al-3Mo-4Sn, 6Al-3Mo-4Zr, and 6Al-3Mo-2Sn-2Zr, were significantly superior at a rupture life of 100 hours, having rupture strengths of 48,000, 45,000 and 52,000 psi, respectively. However, the slopes of the stress-rupture curves were steeper than those for the 6Al-3Mo alloy. As a consequence, at 500 hour rupture life the α complexed alloys with strengths of 35,500, 34,500 and 36,500 psi, as compared to 33,000 psi for the alloy base, were not as superior as indicated by the shorter time values. Extrapolation of the stress-rupture curves suggests that the 6Al-3Mo alloy would be stronger at 1000 hours rupture and beyond.

Although only limited testing was done on the 10Al-3Mo-4Zr alloy at 1020°F and only two tests actually were run to failure, there were definite indications that this alloy has excellent elevated temperature properties. The stress to effect failure in 500 hours was not accurately determined but should be in excess of 50,000 psi.

The 6Al-3Cu, α + compound, alloy was slightly inferior to the 6Al-3Mo alloy at 1020°F.

Comparison of the stress vs. minimum creep rate curves showed that at 800°F the 6Al-1Mo-1Mn-1Cr alloy had a stress of 73,000 psi in common with the base alloy for a creep rate of 0.001%/hour, but was inferior at higher creep rates. At this temperature none of the other β complexed alloys approached the creep resistance of the 6Al-3Mo alloy. The α complexed alloys, 6Al-3Mo-4Sn and 6Al-3Mo-2Sn-2Zr, together with the 6Al-3Cu alloy, had nearly identical creep properties and were significantly more resistant than the 6Al-3Mo alloy (85,000 psi at a creep rate of 0.001%/hour as compared to 73,000 psi for the base). The 6Al-3Mo-4Zr alloy very closely approached the other α complexed alloys but was 2000 psi inferior (83,000 psi at a creep rate of 0.001%/hour). The 10Al-3Mo-4Zr alloy had a creep strength of 97,000 psi for a 0.001%/hour rate.

At 1020°F the β complexed alloys were markedly inferior to the 6Al-3Mo alloy, while the α complexed alloys gave the best performance. Stresses of 49,000 and 28,000 psi correspond to creep rates of 0.1%/hour and 0.01%/hour,

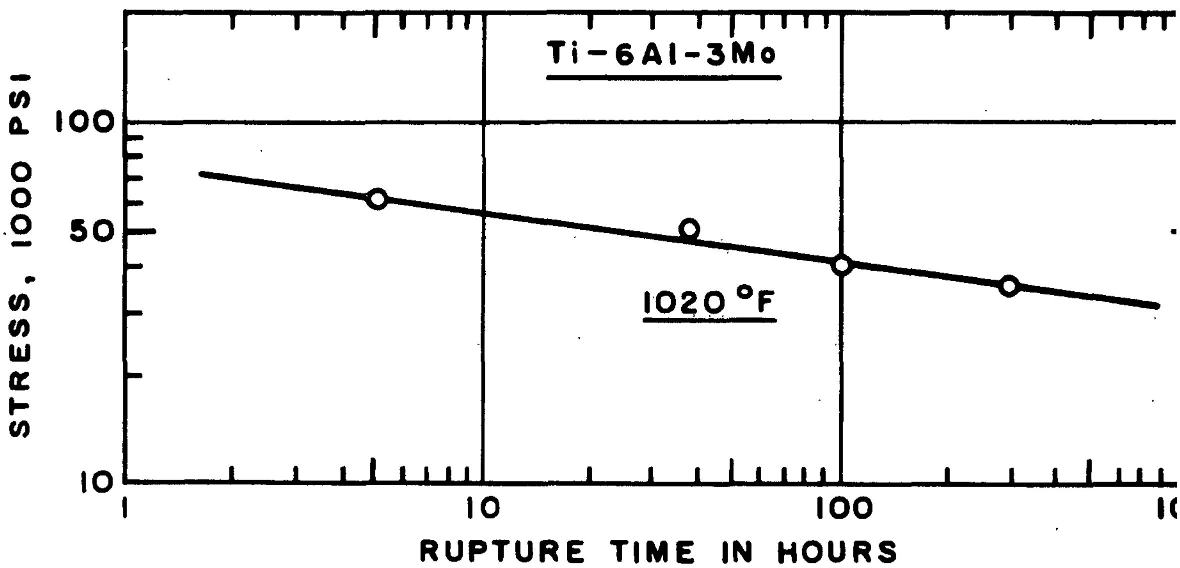


FIG. 20 — STRESS-RUPTURE CURVE FOR ALLOY Ti-6%Al-3%Mo

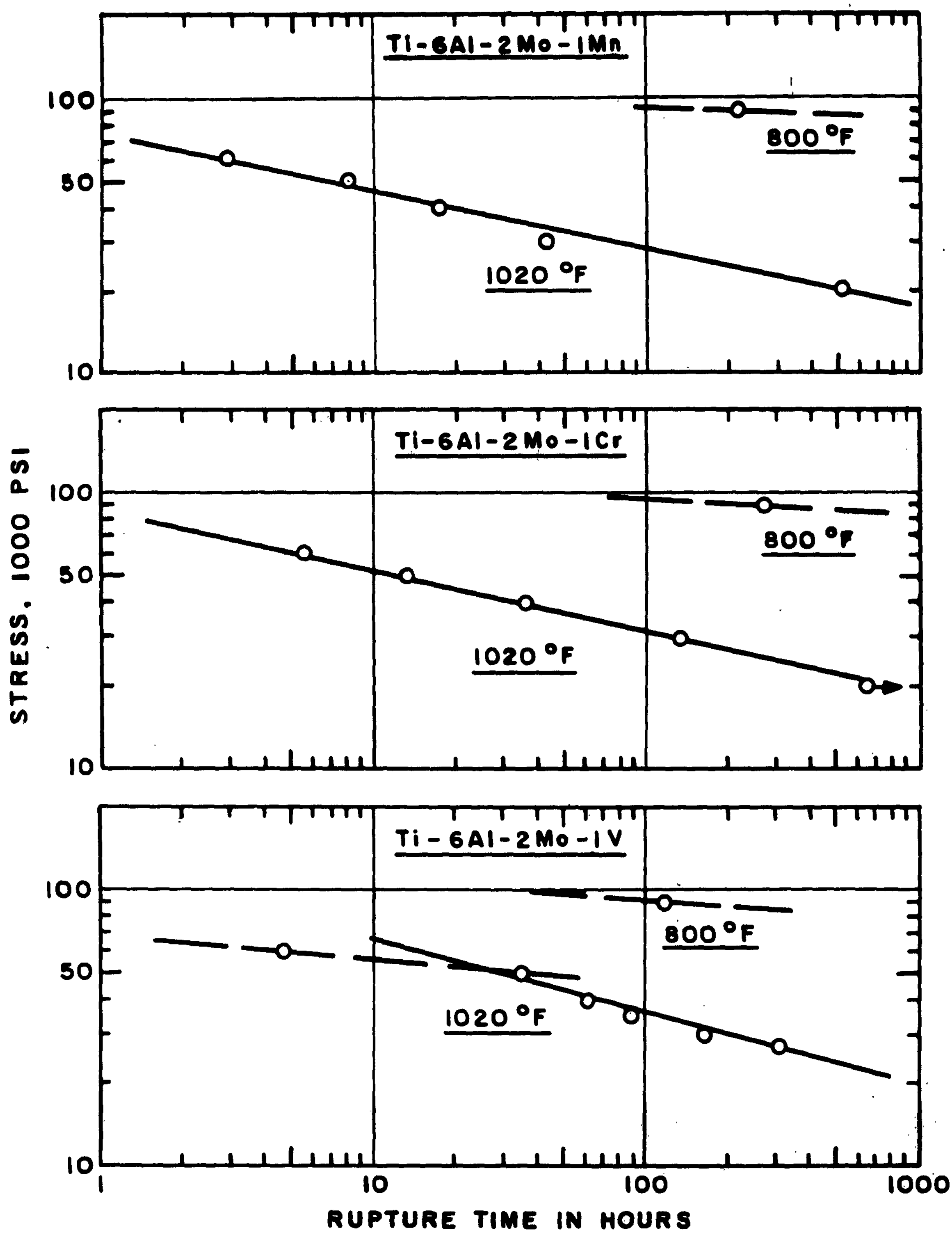


FIG. 21 – STRESS-RUPTURE CURVES FOR ALLOYS Ti-6%Al-2%Mo, WITH 1%Mn, 1%Cr OR 1%V

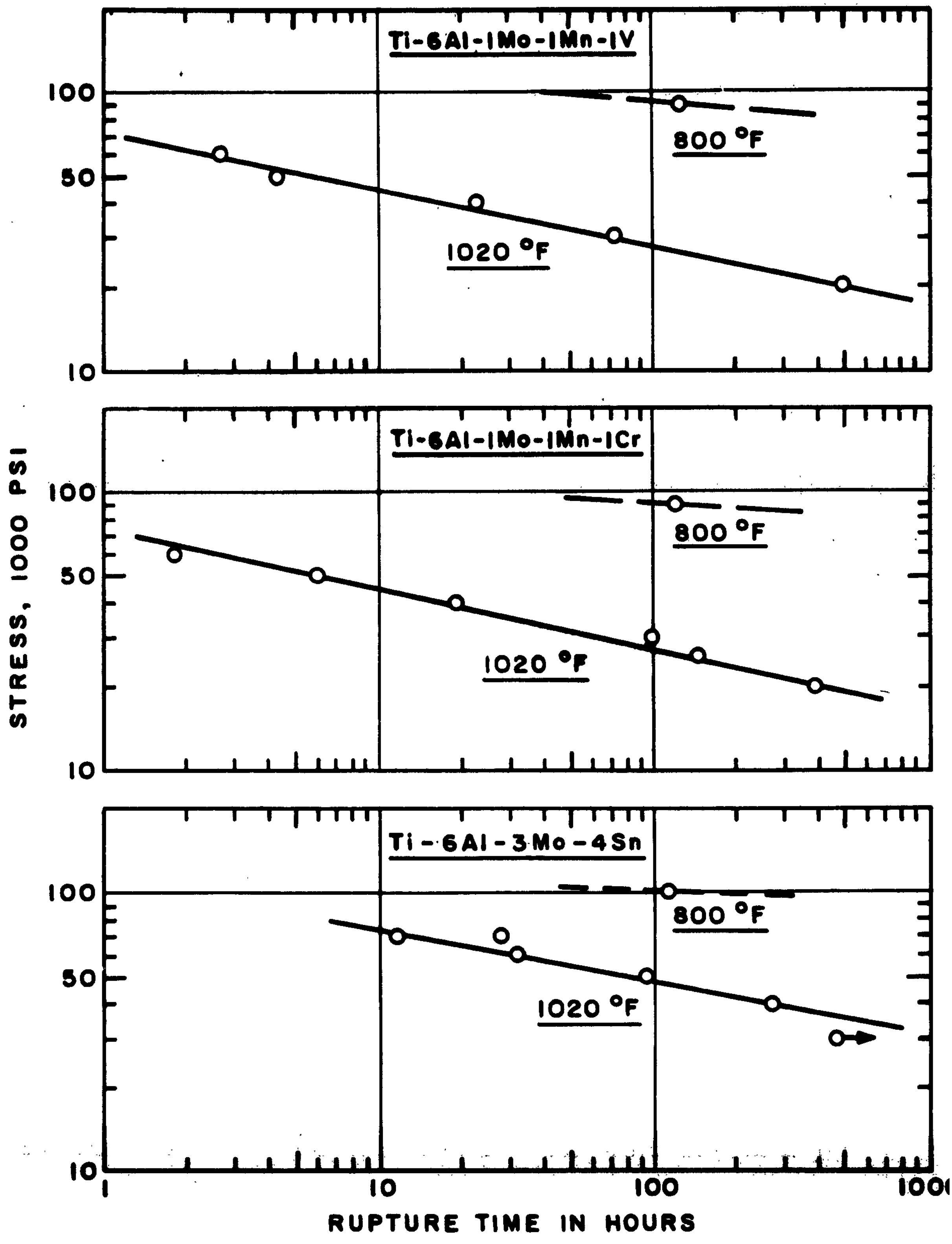


FIG. 22—STRESS-RUPTURE CURVES FOR ALLOYS Ti-6%Al-1%Mo-1%Mn WITH 1%V OR 1%Cr, AND Ti-6%Al-3%Mo-4%Sn

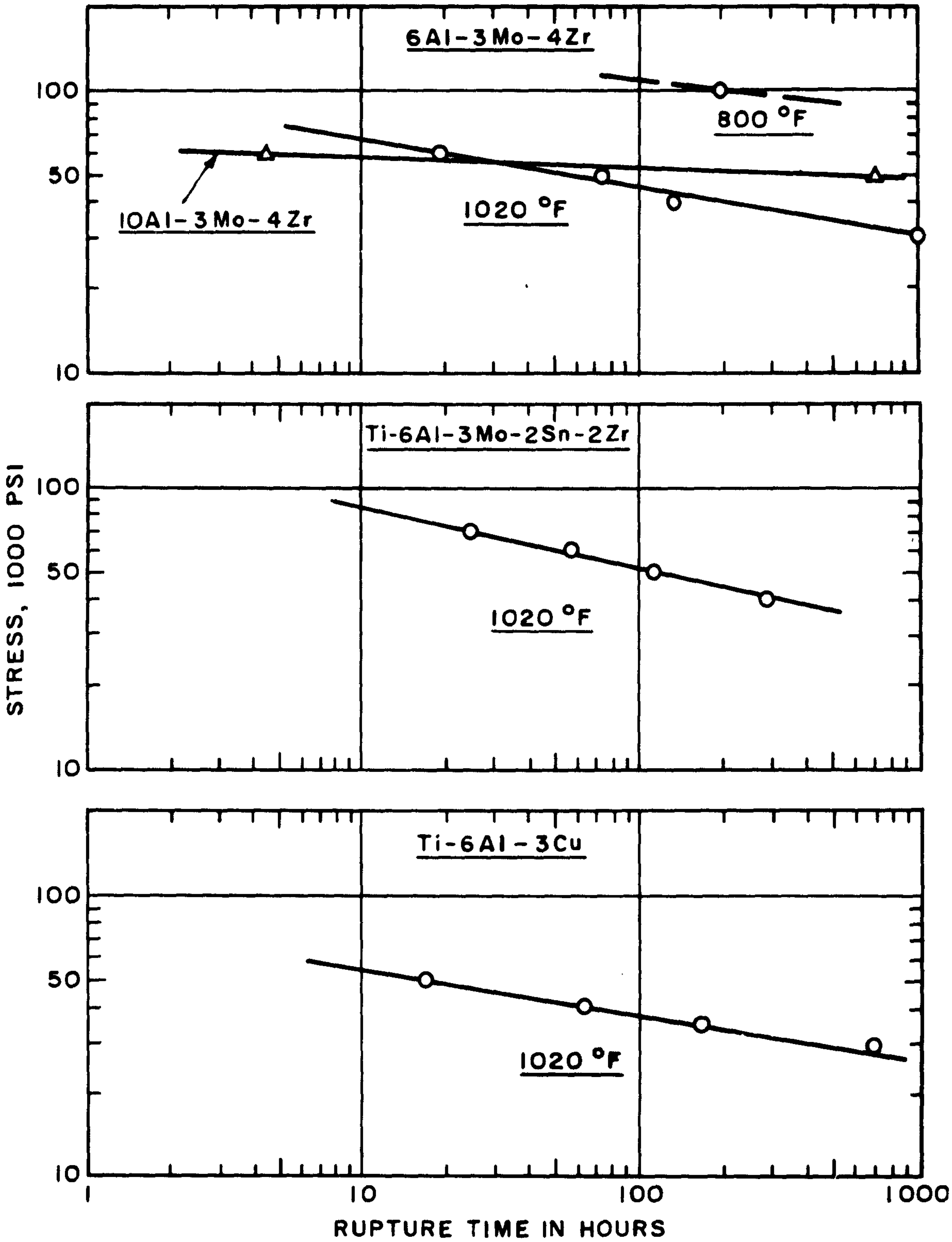


FIG. 23 — STRESS-RUPTURE CURVES FOR ALLOYS Ti-6%Al-3%Mo-4% Zr, Ti-6%Al-3%Mo-2%Sn-2%Zr AND Ti-6%Al-3%Cu

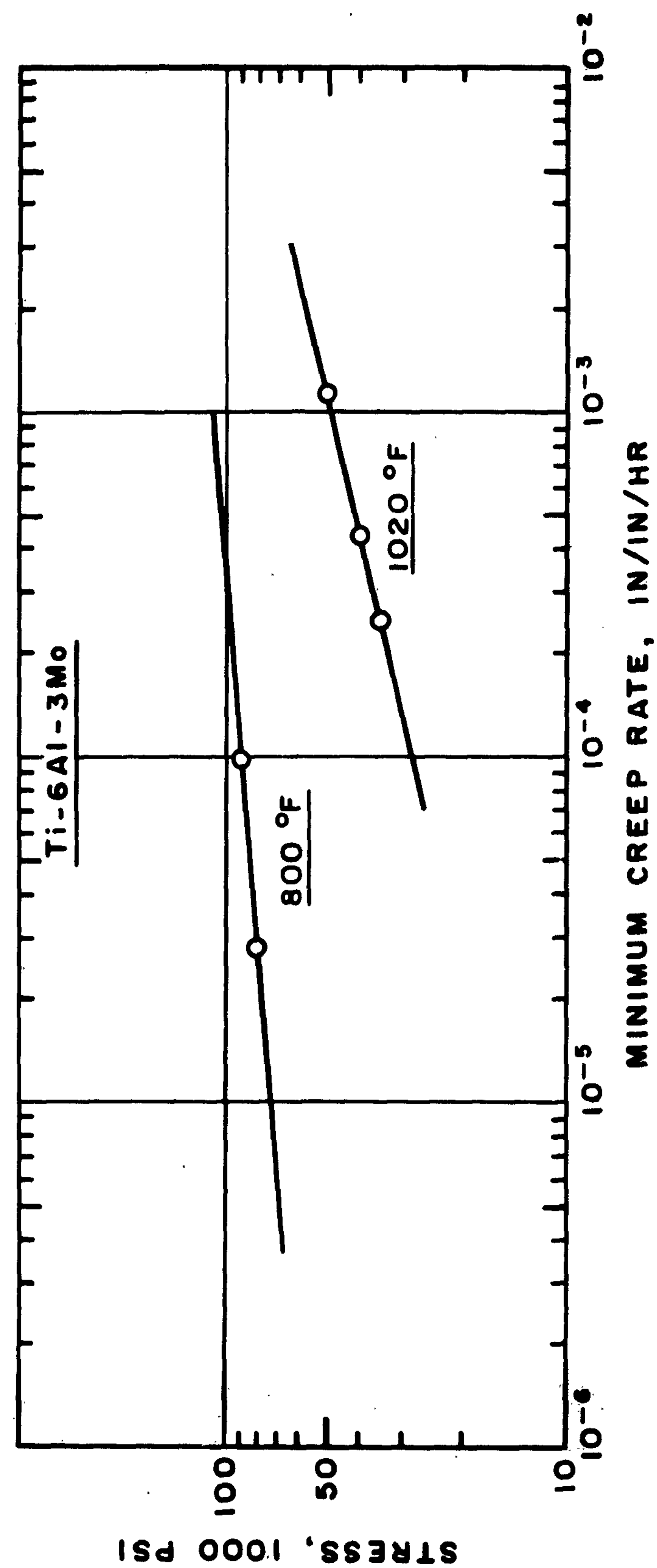


FIG. 24 - MINIMUM CREEP RATE CURVES FOR ALLOY Ti-6% Al-3% Mo

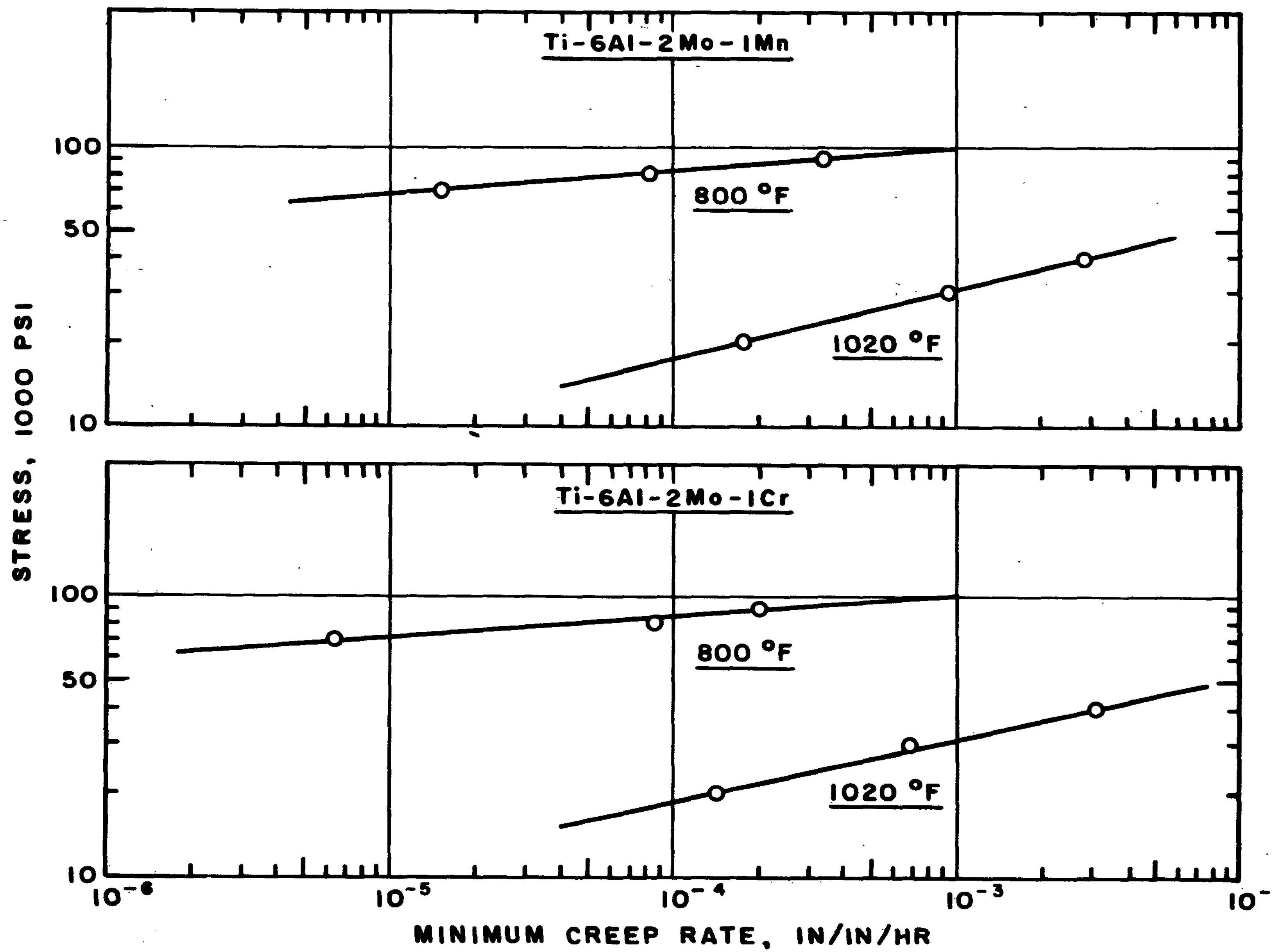


FIG. 25 – MINIMUM CREEP RATE CURVES FOR ALLOYS Ti-6%Al-2%Mo-1%Mn OR 1%Cr.

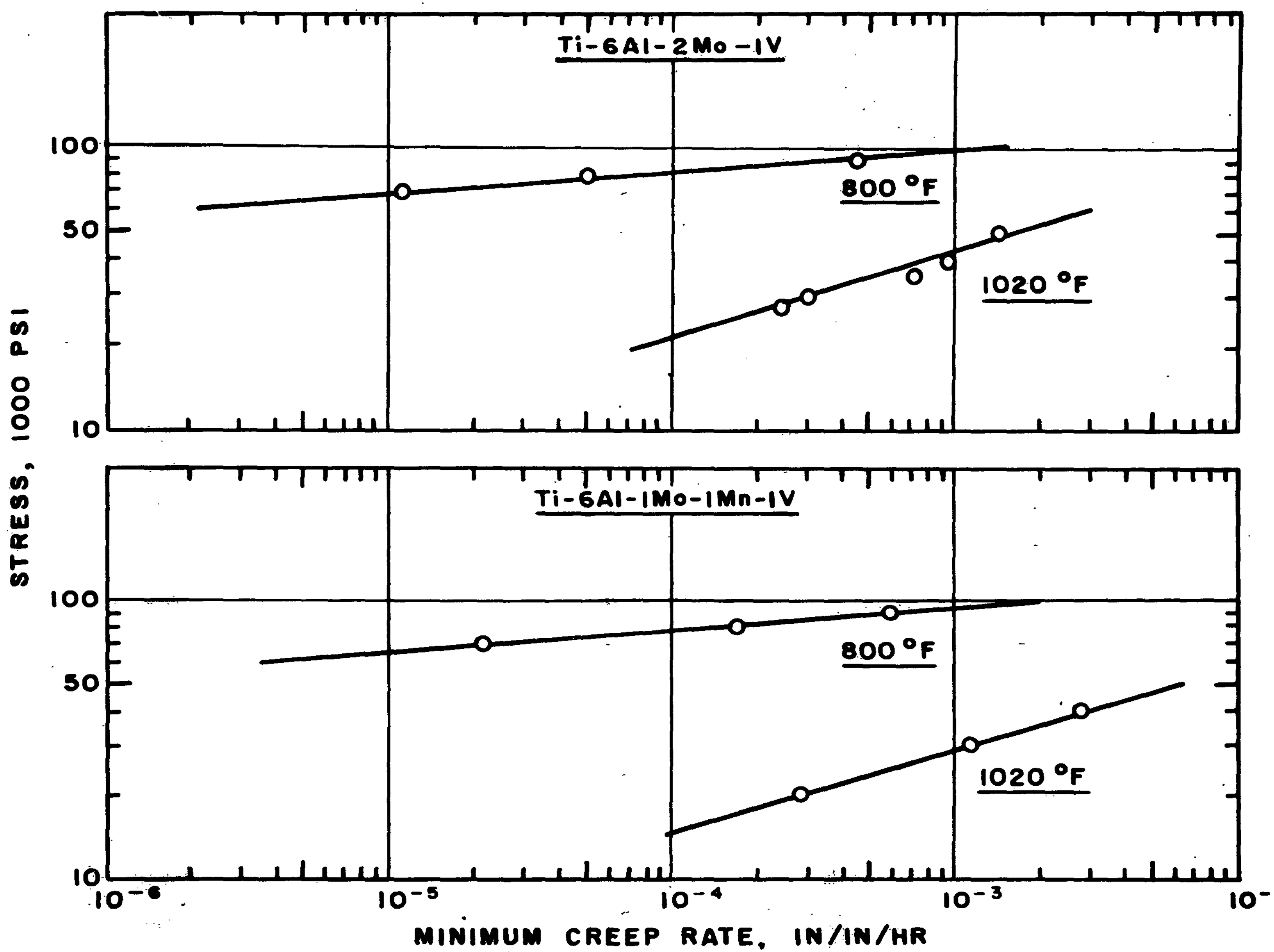


FIG. 26 – MINIMUM CREEP RATE CURVES FOR ALLOYS
Ti-6%Al-2%Mo-1%V AND

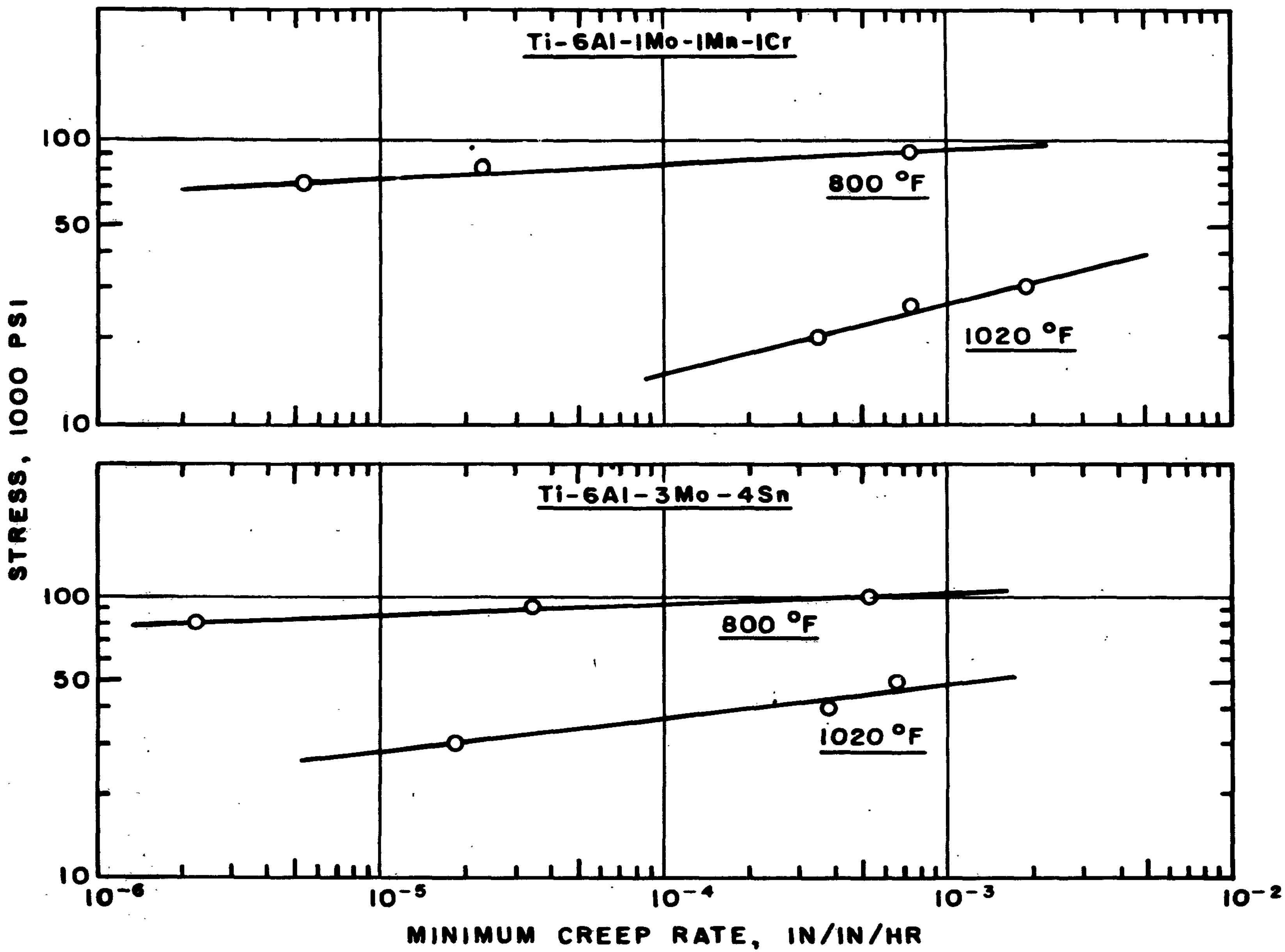


FIG. 27- MINIMUM CREEP RATE CURVES FOR ALLOYS
Ti-6%Al-1%Mo-1%Mn-1%Cr AND
Ti-6%Al-3%Mo-4%Sn

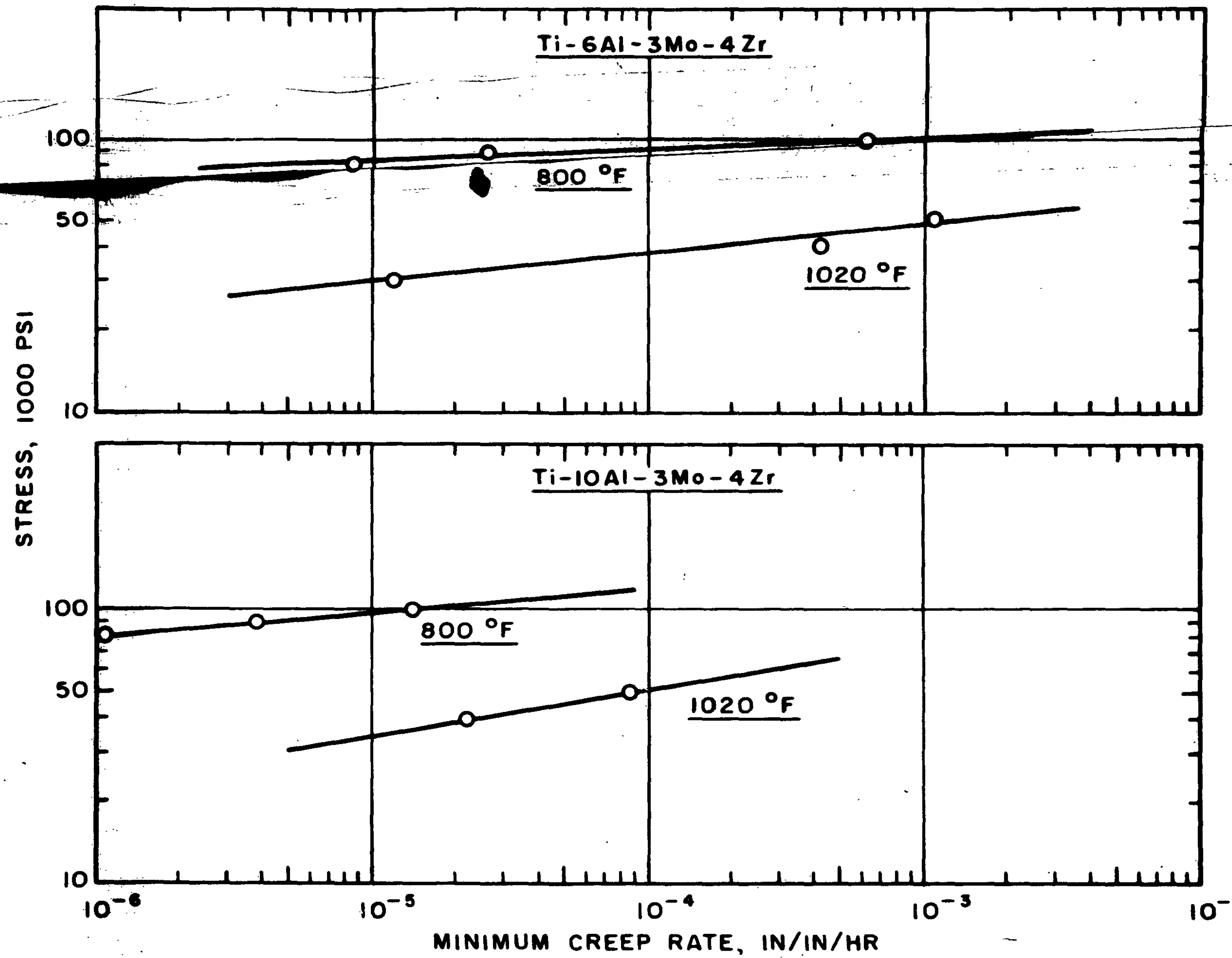


FIG. [illegible] MINIMUM CREEP RATE CURVES FOR ALLOYS

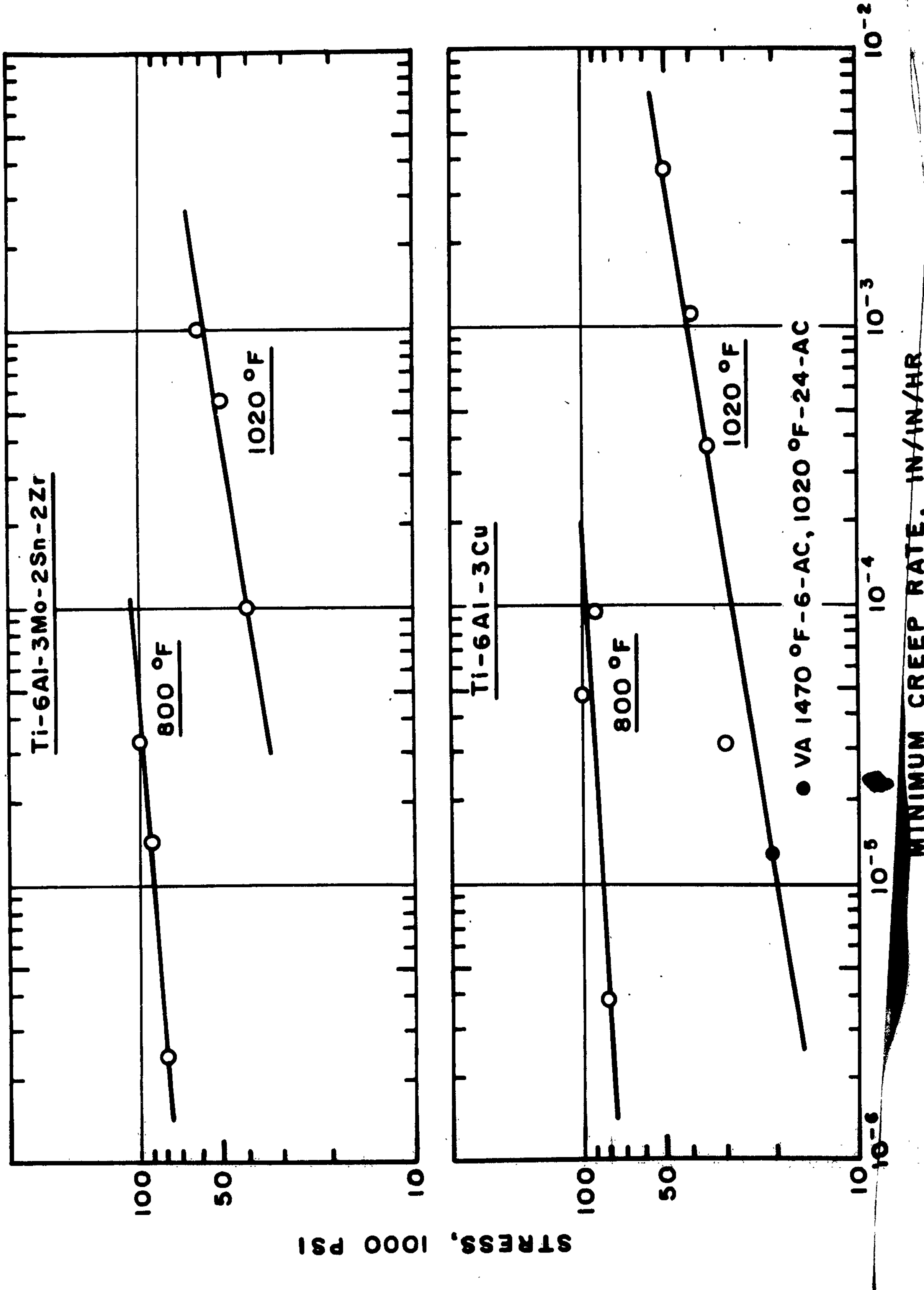


FIG. 29 - MINIMUM CREEP RATE CURVES FOR ALLOYS
Ti-6%Al-3%Mo-2%Sn-2%Zr AND
Ti-6%Al-3%Cu

TABLE 10

CREEP-RUPTURE DATA FOR EXPERIMENTAL FORGING ALLOYS AT 800°F

Alloy*	Stress (psi)	Rupture Time (hrs.)	Primary Creep %	Secondary Creep %	Tertiary Creep %	Total Measured Elongation %	Minimum Creep Rate in/in/hr.
6Al-3Mo	90,000	310.6[a]	2.6	2.9	--	5.5	0.000098
	80,000	304.0[a]	1.6	0.9	--	2.5	0.000028
	70,000	308.8[a]	0.6	0.07	--	b	0.0000029
6Al-2Mo-1Mn	90,000	216.5	1.8	3.7	15.5	21.0	0.00034
	80,000	306.4[a]	0.5	2.5	--	3.0	0.000082
	70,000	307.0[a]	0.5	0.5	--	1.0	0.000015
6Al-2Mo-1Cr	90,000	270.1	2.5	2.8	14.2	19.5	0.00020
	80,000	304.7[a]	2.4	2.6	--	5.0	0.000086
	70,000	305.7[a]	0.4	0.1	--	b	0.0000064
6Al-2Mo-1V	90,000	119.6	4.2	2.6	14.7	21.5	0.00045
	80,000	303.6[a]	1.7	1.3	--	3.0	0.000051
	70,000	307.2[a]	0.8	0.2	--	1.0	0.000011
6Al-1Mo-1Mn-1V	90,000	126.7	2.5	4.0	8.5	15.0	0.00059
	80,000	304.9[a]	2.3	5.2	--	7.5	0.00017
	70,000	311.0[a]	0.3	0.7	--	1.0	0.000022
6Al-1Mo-1Mn-1Cr	90,000	119.8	2.5	6.6	11.4	20.5	0.00073
	80,000	308.0[a]	1.6	0.4	--	2.0	0.000023
	70,000	307.2[a]	0.9	0.1	--	1.0	0.0000053
6Al-3Mo-4Sn	100,000	112.9	5.0	3.6	5.4	14.0	0.00053
	90,000	304.1[a]	0.1	0.9	--	1.0	0.000034
	80,000	306.1[a]	---	0.07	--	b	0.0000022
6Al-3Mo-4Zr	100,000	194.7	4.1	8.0	7.9	20.0	0.00061
	90,000	319.7[a]	0.3	0.7	--	1.0	0.000026

TABLE 10 (continued)

CREEP-RUPTURE DATA FOR EXPERIMENTAL FORGING ALLOYS AT 800°F

Alloy*	Stress (psi)	Rupture Time (hrs.)	Primary Creep %	Secondary Creep %	Tertiary Creep %	Total Measured Elongation %	Minimum Creep Rate in/in/hr.
10Al-3Mo-4Zr	100,000	332.8[a]	---	0.4	--	b	0.000014
	90,000	306.2[a]	---	0.1	--	b	0.0000038
	80,000	312.4[a]	---	0.03	--	b	0.0000011
6Al-3Mo-2Sn-2Zr	100,000	328.2[a]	1.1	0.9	--	2.0	0.000033
	90,000	311.1[a]	---	0.4	--	b	0.000015
	80,000	307.9[a]	---	0.07	--	b	0.0000025
6Al-3Cu	100,000	304.0[a]	1.0	1.3	3.2	5.5	0.000048
	90,000	304.3[a]	0.4	1.9	1.2	3.5	0.000095
	80,000	331.4[a]	---	0.1	--	b	0.0000038

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

a Test stopped at time indicated. No failure.

b No measurable extension.

TABLE 11

CREEP-RUPTURE DATA FOR EXPERIMENTAL FORGING ALLOYS AT 1020°F

Alloy*	Stress (psi)	Rupture Time (hrs.)	Primary Creep %	Secondary Creep %	Tertiary Creep %	Total Measured Elongation %	Minimum Creep Rate in/in/hr.
6Al-3Mo	60,000	5.0	1.0	0.0	4.5	5.5	---
	50,000	37.9	0.6	1.0	15.9	17.5	0.0011
	40,000	100.1	0.2	1.8	24.0	26.0	0.00044
	35,000	295.8	0.08	3.2	18.7	22.0	0.00025
6Al-2Mo-1Mn	60,000	2.9	1.4	0.0	26.6	28.0	---
	50,000	8.0	0.8	0.0	28.2	29.0	---
	40,000	17.2	0.6	1.0	22.4	24.0	0.0027
	30,000	43.4	0.5	1.0	31.5	33.0	0.00093
	20,000	521.3	0.2	4.5	24.8	29.5	0.00018
6Al-2Mo-1Cr	60,000	5.5	---	0.0	--	24.0	
	50,000	13.1	---	0.0	--	15.5	
	40,000	36.0	0.6	0.3	28.6	29.5	0.0031
	30,000	134.0	0.2	4.6	27.2	32.0	0.00068
	20,000	644.2[a]	0.1	3.2	9.7	13.0	0.00014
6Al-2Mo-1V	60,000	4.7	0.6	0.0	16.4	17.0	---
	50,000	35.1	0.4	2.5	10.1	13.0	0.0014
	40,000	62.0	0.6	2.0	14.9	17.5	0.00095
	35,000	88.9	0.3	3.0	20.2	23.5	0.00072
	30,000	167.2	0.25	1.6	22.15	24.0	0.00030
	27,500	316.1	?	1.6	~54	56.0	0.00025
6Al-1Mo-1Mn-1V	60,000	2.7	0.4	0.0	24.1	24.5	---
	50,000	4.3	0.5	0.0	31.0	31.5	---
	40,000	22.9	0.3	1.6	14.1	16.0	0.0028
	30,000	71.5	?	2.4	~24	27.0	0.0011
	20,000	493.6	0.3	8.0	63.7	72.0	0.00029

TABLE 11 (continued)

CREEP-RUPTURE DATA FOR EXPERIMENTAL FORGING ALLOYS AT 1020°F

Alloy*	Stress (psi)	Rupture Time (hrs.)	Primary Creep %	Secondary Creep %	Tertiary Creep %	Total Measured Elongation %	Minimum Creep Rate in/in/hr.
6Al-1Mo-1Mn-1Cr	60,000	1.8	?	0.0	~21	22.0	---
	50,000	6.0	1.0	0.0	31.0	32.0	---
	40,000	19.2	0.3	0.0	27.7	28.0	---
	30,000	98.4	0.4	1.5	34.6	36.5	0.0019
	26,000	146.8	0.2	2.7	56.1	59.0	0.00075
	20,000	383.7	0.2	5.6	39.2	45.0	0.00035
6Al-3Mo-4Sn	70,000	11.7	0.7	0.0	12.3	13.0	---
	60,000	31.4	0.5	0.0	12.0	12.5	---
	50,000	93.2	0.6	3.0	12.0	15.0	0.00067
	40,000	269.8	0.5	5.0	9.0	14.5	0.00038
	30,000	456.7[a]	0.3	0.7	--	1.0	0.000019
6Al-3Mo-4Zr	60,000	19.3	1.1	0.0	21.4	22.5	---
	50,000	73.8	0.6	1.6	17.3	19.5	0.0011
	40,000	130.9	0.3	3.0	37.7	41.0	0.00043
	30,000	1061.2	b	10.9	22.1	33.0	0.000012
10Al-3Mo-4Zr	60,000	4.5	0.6	0.0	19.4	20.0	---
	50,000	705.3	0.5	2.5	9.0	12.0	0.000086
	40,000	1001.3[a]	0.2?	2.2	--	3.0	0.000022
6Al-3Mo-2Sn-2Zr	70,000	24.8	?	0.0	~21	22.0	---
	60,000	57.0	0.3	2.4	22.3	25.0	0.0010
	50,000	114.2	?	3.0	~22	25.0	0.00056
	40,000	295.1	?	1.5	~53	55.0	0.00010
6Al-3Cu	50,000	17.0	0.5	2.5	19.0	22.0	0.0037
	40,000	62.3	0.4	3.5	18.1	22.0	0.0012
	35,000	168.0	0.2	2.5	16.3	19.0	0.00037
	30,000	715.6	0.2	1.5	5.3	7.0[c]	0.000032
	10,000	307.2[a]	---	--	--	1.0	0.000013

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.
a Test stopped at time indicated. No failure.
b No measurable extension.
c Failure occurred in gage mark.

TABLE 12

SUMMARY OF RUPTURE STRENGTH

AND CREEP RESISTANCE OF EXPERIMENTAL FORGING ALLOYS

	Stress to Rupture at 1020°F			Stress to Give Indicated Creep Rate				
				800°F			1020°F	
Alloy	10 hrs	100 hrs	500 hrs	0.1%/hr	0.01%/hr	0.001%/hr	0.1%/hr	0.01%/hr
6Al-3Mo	55,500	40,000	33,000	---	90,000	73,000	49,000	28,000
6Al-2Mo-1Mn	46,000	28,000	20,000	99,000	82,000	68,000	31,000	17,500
6Al-2Mo-1Cr	52,000	32,000	22,000	100,000	84,000	72,000	31,000	18,500
6Al-2Mo-1V	----	35,000	21,500	98,000	82,000	68,000	43,500	21,000
6Al-1Mo-1Mn-1V	45,000	27,500	20,000	94,000	78,000	64,500	29,000	14,500
6Al-1Mo-1Mn-1Cr	45,000	27,000	19,000	92,000	82,000	73,000	26,000	15,000
6Al-3Mo-4Sn	73,000	48,000	35,500	102,000	94,000	85,000	48,000	37,000
6Al-3Mo-4Zr	67,000	45,000	34,500	102,000	92,000	83,000	48,000	37,000
10Al-3Mo-4Zr	----	----	>50,000	---	----	97,000	----	51,000
6Al-3Mo-2Sn-2Zr	84,000	52,000	36,500	---	96,000	85,000	57,000	40,000
6Al-3Cu	54,000	38,000	31,000	---	96,000	85,000	41,000	28,500

respectively, for the 6Al-3Mo alloy. At the higher rate the 6Al-3Mo-4Sn and 6Al-3Mo-4Zr alloys were similar at a stress of 48,000 psi, but were more resistant at the 0.01%/hour rate at a stress of 37,000 psi.

The 6Al-3Mo-2Sn-2Zr alloy was most creep resistant, requiring stresses of 57,000 and 40,000 psi for creep rates of 0.1%/hour and 0.01%/hour, respectively. It is interesting to note that this alloy at 1020°F is weaker than the 7Al-3Mo (1K) alloy(3) on a stress-rupture comparison (52,000 and 36,500 psi for 100 and 500 hour life, respectively, for the 6Al-3Mo-2Sn-2Zr alloy, as compared to stresses for the 7Al-3Mo alloy of 54,500 and 46,000 psi) but is identical on the basis of creep resistance (57,000 and 40,000 psi for rates of 0.1%/hour and 0.01%/hour, respectively).

The 10Al-3Mo-4Zr alloy at 1020°F had a creep strength for a rate of 0.01%/hour of approximately 51,000 psi.

The 6Al-3Cu alloy had inferior creep resistance to the 6Al-3Mo alloy at a creep rate of 0.1%/hour (41,000 psi) but was comparable at a rate of 0.01%/hour (28,000 psi).

It is apparent from the above findings that in complexing of the α phase with tin and zirconium lies the hope for alloy improvement. The elevated temperature performance of the 10Al-3Mo-4Zr alloy is noteworthy. Its potential applicability is hampered by its poor room temperature properties, however. Because of the potential exhibited by this alloy and the 6Al-3Cu alloy, an attempt to improve their room temperature properties would be amply justified.

3. Stability Characteristics

The results of room temperature tensile tests conducted on specimens tested initially in creep for periods of 300 or 1000 hours are summarized in Table 13. The base 6Al-3Mo alloy was not significantly affected by the exposure conditions. The 6Al-2Mo-1Mn alloy which was low in ductility initially was also unaffected. (One specimen was brittle as a result of an imperfection.) Two specimens of the 6Al-2Mo-1Cr alloy (800°F-70,000 psi-305.7 hours and 1020°F-20,000 psi-644.2 hours) were noticeably influenced by stress-aging. The lack of ductility in the specimen exposed at 1020°F is perhaps explainable by the fact that this test piece extended 13% in creep. Tests on the 6Al-2Mo-1V alloy were variable; exposure at 800°F under a stress of 80,000 psi showed small change in properties, whereas exposure under a stress of 70,000 psi brought about a considerable drop in ductility.

The 6Al-1Mo-1Mn-1V and 6Al-1Mo-1Mn-1Cr alloys also lost ductility through exposure. The 1V bearing alloy aged under a stress of 80,000 psi at 800°F was completely brittle.

The α complexed alloys were also seriously affected by stress-aging. Again, the loss in ductility was not systematic and no attempt is made to explain the behavior.

As in the case of unexposed specimens, 6Al-3Cu alloy specimens which were exposed to stress at 800° and 1020°F had poor or good ductility depending upon the microstructure. The two specimens with good ductility were equiaxed; while the others were Widmanstätten.

TABLE 13

STABILITY OF ROOM TEMPERATURE PROPERTIES OF FORGING ALLOYS TO ELEVATED TEMPERATURE EXPOSURE

Alloy	Heat Treatment	Exposure Conditions: Temp (°F)	Stress (psi)	Time (hrs)	Def. (in.)	Before Exposure: UTS (psi)	% RA	% El	After Exposure: UTS (psi)	% RA	% El
6Al-3Mo	1	800	90,000	310.6	0.055	145,000	28.5	12.5	150,500	25.0	11.0
	1	800	80,000	304.0	0.025	145,000	28.5	12.5	146,800	20.0	13.5
6Al-2Mo-1Mn	1	800	80,000	306.4	0.030	166,700	8.5	3.5	155,800	8.0	3.5
	1	800	70,000	307.0	0.010	166,700	8.5	3.5	79,300	0.0[a]	0.0[a]
	2	800	70,000	310.4	0.010	156,200	23.0	13.0	153,500	27.0	15.0
	2	1020	5,000	309.0	0.005	156,200	23.0	13.0	151,000	18.5	11.0
	3	1020	5,000	93.0	0.010	149,000	23.0	12.0	141,500	19.0	8.5
6Al-2Mo-1Cr	1	800	80,000	304.7	0.050	147,500	30.0	14.0	152,200	26.0	15.0
	1	800	70,000	305.7	0.007	147,500	30.0	14.0	148,200	15.5	10.0
	1	1020	20,000	644.2	0.130	147,500	30.0	14.0	148,000	15.5	5.5
	1	1020	7,500	1000.6	0.005	147,500	30.0	14.0	145,000	29.0	13.0
6Al-2Mo-1V	1	800	80,000	303.6	0.030	144,500	40.0	16.0	149,000	32.5	16.0
	1	800	70,000	307.2	0.010	144,500	40.0	16.0	166,000	10.5	8.5
	2	800	70,000	306.3	0.010	137,500	34.0	13.0	144,000	38.0	15.0
6Al-1Mo-1Mn-1Cr	1	800	80,000	308.0	0.020	153,500	31.5	14.0	159,800	16.0	11.0
	1	800	70,000	307.2	0.010	153,500	31.5	14.0	153,200	8.0	4.5
	1	1020	7,500	1006.1	0.005	153,500	31.5	14.0	144,500	5.5	3.0
6Al-1Mo-1Mn-1V	1	800	80,000	304.9	0.075	156,500	44.0	17.5	161,000	0.0	0.0
	1	800	70,000	311.0	0.010	156,500	44.0	17.5	153,500	8.0	2.5
6Al-3Mo-4Sn	1	800	90,000	304.1	0.010	156,000	23.0	14.5	168,300	11.0	7.0
	1	800	80,000	306.1	0.002	156,000	23.0	14.5	169,000	19.5	14.0
	1	1020	30,000	456.7	0.010	156,000	23.0	14.5	58,600[b]	--	0.0
	2	1020	30,000	320.0	0.015	155,500	23.0	14.0	156,000	8.5	5.5

TABLE 13 (continued)

STABILITY OF ROOM TEMPERATURE PROPERTIES

OF FORGING ALLOYS TO ELEVATED TEMPERATURE EXPOSURE

Alloy	Heat Treatment	Exposure Conditions Temp (°F)	Stress (psi)	Time (hrs)	Def. (in.)	Before Exposure UTS (psi)	% RA	% El	After Exposure UTS (psi)	% RA	% El
6Al-3Mo-4Zr	1	800	90,000	319.7	0.010	150,000	28.0	12.0	160,000	25.0	13.0
	1	800	80,000	300.2	0.010	150,000	28.0	12.0	151,000	29.0	16.0
6Al-3Mo-2Sn-2Zr	1	800	100,000	328.2	0.020	156,000	26.5	13.0	165,200	9.5	4.0
	1	800	90,000	311.1	0.005	156,000	26.5	13.0	158,200	6.5	5.5
	2	800	90,000	309.5	0.025	164,300	25.0	14.0	162,000	14.0	11.5
	3	800	90,000	317.5	0.010	164,300	25.0	14.0	163,000	14.5	11.0
	1	800	80,000	307.9	0.001	156,000	26.5	13.0	164,800	22.5	12.5
6Al-3Cu	1	800	100,000	304.0	0.055	147,000	5.5	5.0	82,800	c	c
	1	800	90,000	304.3	0.035	147,000	5.5	5.0	138,000	c	c
	1	800	80,000	331.4	0.002	147,000	5.5	5.0	105,200	1.5	1.0
	2	800	80,000	300.2	0.015	148,000	2.5	2.0	---	4.5	3.5
	1	1020	5,000	1000.5	0.005	147,000	5.5	5.0	110,500	2.5	c
	3	1020	5,000	297.0	0.035	151,000	23.0	11.0	144,000	17.0	9.5
	2	1020	10,000	307.2	0.010	148,000	2.5	2.0	---	19.5	10.5

1 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

2 VA 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

3 VA 1560°F-4 hrs-AC, 1020°F-24 hrs-AC.

a Forging seam.

b Fracture occurred in threads.

c Fracture occurred outside gage marks.

C. Results for Sheet Alloys

1. Tensile Properties

The tensile characteristics of three sheet alloys, Ti-6Al-2V-1Mo, Ti-6Al-2V-2Mo, and Ti-6Al-3V-1Mo, were determined at room temperature, 620°, 800° and 1020°F. Duplicate tests were run at each test temperature. Specimens were given the same heat treatment given the forging alloys, namely, 1470°F-6 hours-air cool, 1020°F-24 hours-air cool. The test data are included as Table 14 and plotted in Figures 30 to 32.

Of the three modifications of the 6Al-4V basic composition there was little to choose between the 6Al-2V-2Mo and 6Al-3V-1Mo compositions at the lower temperatures. The 2Mo bearing alloy was slightly stronger at 1020°F. The 6Al-2V-1Mo alloy was about 10,000 psi weaker than the other two modifications over the entire test temperature range. On a ductility basis all three alloys were quite comparable. It is interesting to note that ductility did not change appreciably with temperature.

2. Creep-Rupture Properties

Prior to the initiation of the sheet test program it was believed that the extensometers employed in creep testing of bar specimens could be modified for similar testing of sheet specimens. It was planned to attach the extensometer to the specimen holders and compensate for the creep occurring outside the 1 in. gage length. Because of excessive elongation of the test pieces at 1020°F, coupled with slippage of the specimen within the holders, the proposed creep measuring procedure was abandoned. As a consequence, extension with time data are not available for the sheet alloys. At 800°F the test specimens were gage marked very accurately (1.0000 in. ± 0.0002 in.) and in the instances where rupture did not occur the reported total elongation values are correct within ±0.0002 in. Tests at 1020°F were run to failure and total elongation values for these specimens and all specimens at 800°F which ruptured are accurate to within ±0.005 in. The data are recorded in Table 15 and curves of stress versus rupture life are presented for the three alloys at 1020°F in Figure 33.

On the basis of rupture strength at 1020°F and the limited total extension data at 800°F the 6Al-2V-2Mo alloy appears to be the best modification of the three. The 100 and 500 rupture strengths of this material, 24,000 and 15,000 psi, respectively, are well below values for the 6Al-4V alloy for similar rupture lives, namely, 33,000 and 26,000 psi. The difference in strength levels may be tempered somewhat by the fact that the 6Al-4V alloy was tested in bar form. Test piece geometry probably plays a significant part in the wide spread in strength.

3. Stability Characteristics

Specimens for stability studies were taken from unbroken test pieces creep tested at 800°F. Only four such specimens were available. The 6Al-2V-1Mo alloy tested: 800°F-70,000 psi-306.2 hours had an ultimate room temperature strength of 136,000 psi as compared to an unexposed ultimate of 141,000 psi. Ductility was lowered from 11 to 8%. The 6Al-2V-2Mo alloy creep tested:

TABLE 14

TENSILE DATA FOR EXPERIMENTAL SHEET ALLOYS

Alloy*	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Elongation %	Modulus of Elasticity (psi x 10^{-6})	DPH (20 Kg. Load)
			Room Temperature		
6Al-2V-1Mo	140,200	---	11.0	--	330
	142,000	131,000	11.0	16.8	328
6Al-2V-2Mo	153,700	144,000	11.0	16.8	357
	151,600	144,000	11.0	17.5	347
6Al-3V-1Mo	153,000	150,000	a	12.9	350
	151,600	149,500	11.0	15.6	359
			620°F		
6Al-2V-1Mo	98,500	82,800	11.5	15.5	
	102,200	85,500	11.5	11.7	
6Al-2V-2Mo	113,000	97,700	12.5	12.1	
	109,600	95,500	12.0	12.1	
6Al-3V-1Mo	113,500	100,000	13.0	11.5	
	107,000	92,000	12.0	15.9	
			800°F		
6Al-2V-1Mo	93,200	80,200	12.0	6.4	
	93,800	79,000	11.0	8.6	
6Al-2V-2Mo	106,000	79,300	12.0	--	
	102,000	85,700	12.5	6.8	
6Al-3V-1Mo	104,000	76,400	10.0	7.2	
	104,200	77,500	12.0	5.5	
			1020°F		
6Al-2V-1Mo	79,500	64,400	16.0	8.9	
	82,000	67,000	14.0	8.9	
6Al-2V-2Mo	90,000	74,400	16.5	7.8	
	93,200	83,700	13.5	6.6	
6Al-3V-1Mo	85,100	69,600	20.0	9.0	
	87,800	75,400	19.5	--	

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.
a Failure occurred outside gage marks.

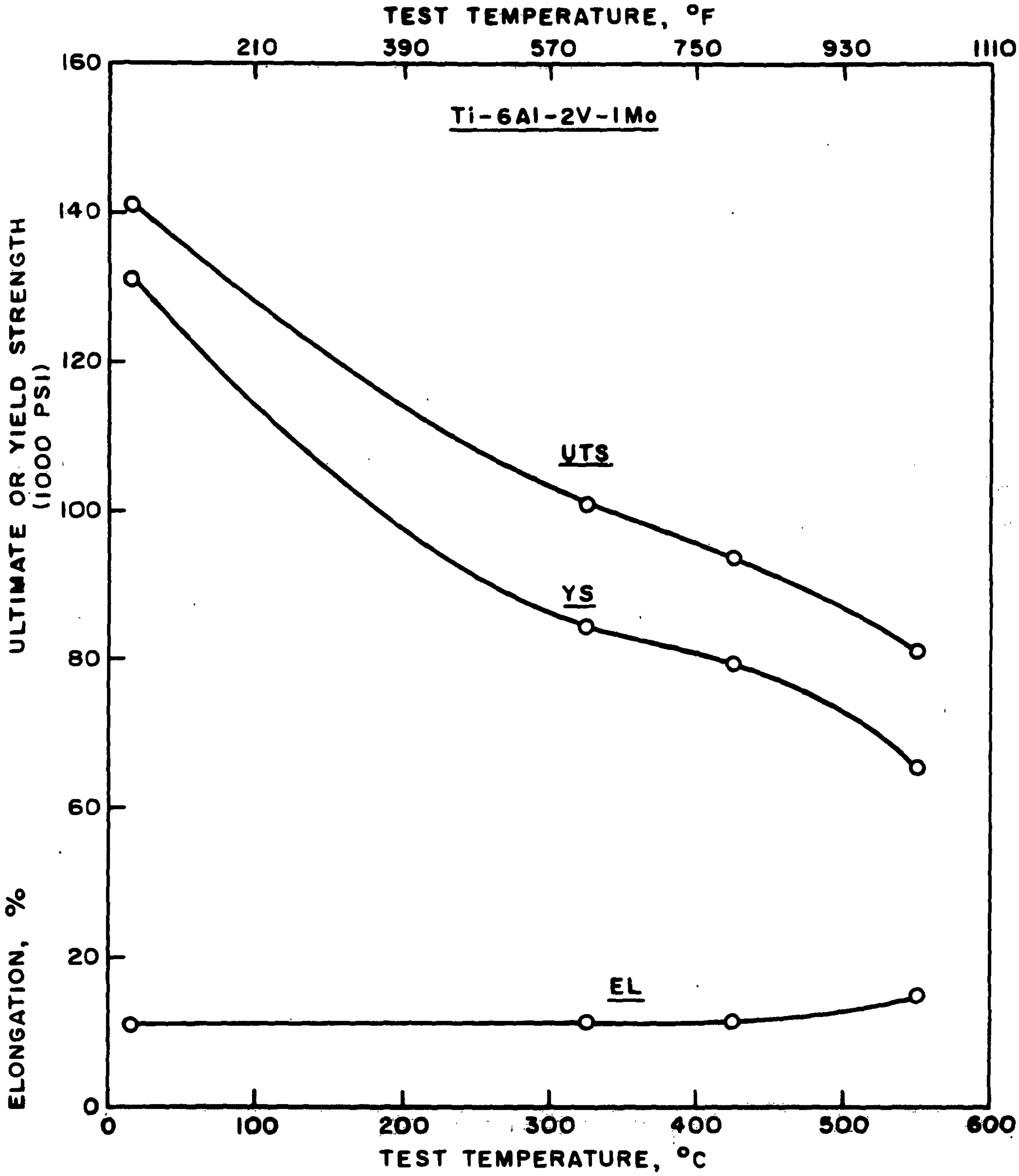


FIG. 30 – TENSILE PROPERTIES OF Ti-6%Al-2%V-1%Mo ALLOY

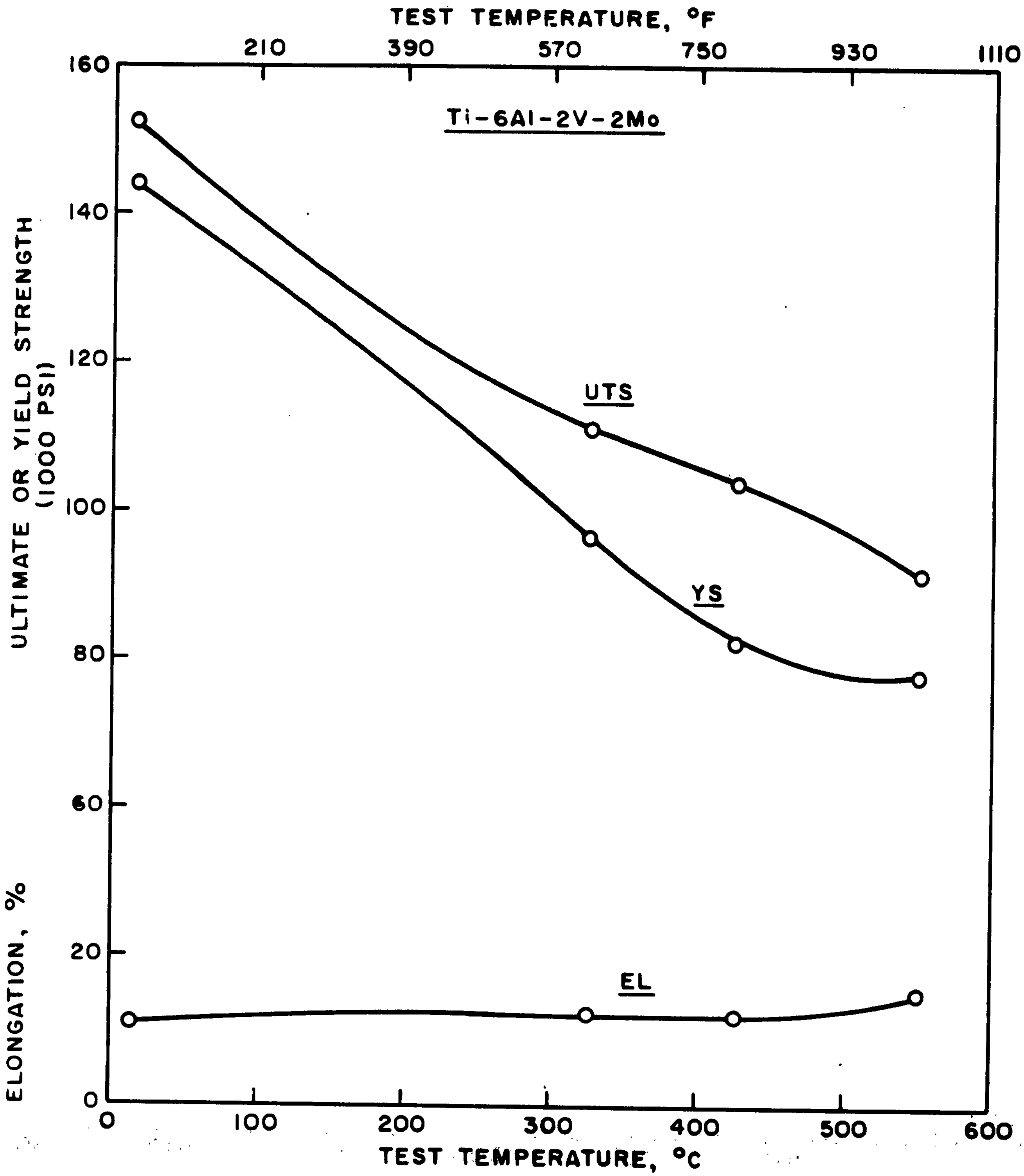


FIG. 31 – TENSILE PROPERTIES OF Ti-6%Al-2%V-2%Mo ALLOY

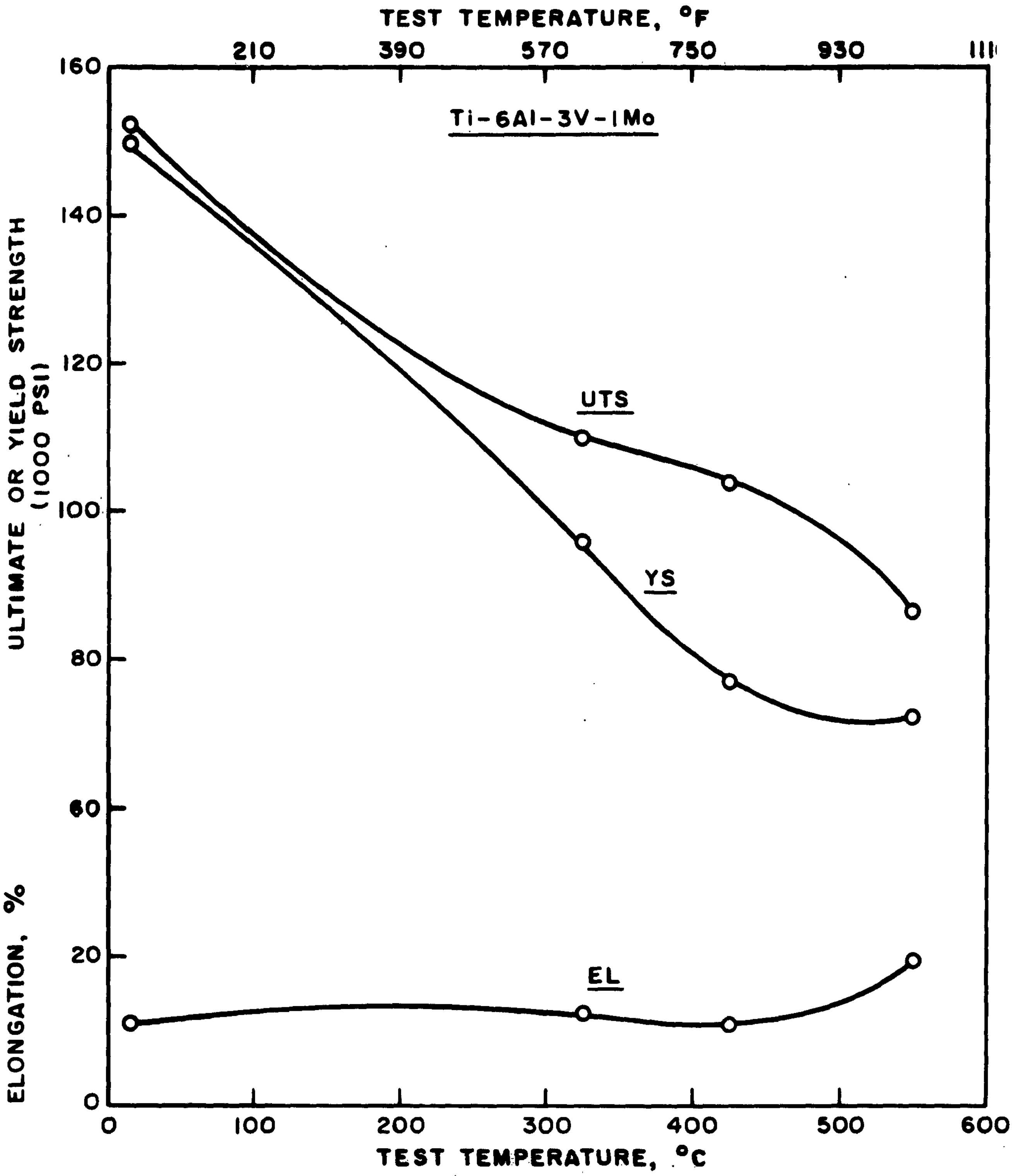


FIG. 32 - TENSILE PROPERTIES OF Ti-6%Al-3%V-1%Mo ALLOY

TABLE 15

CREEP-RUPTURE DATA FOR SHEET ALLOYS AT 800° AND 1020°F

Alloy*	Test Temp (°F)	Stress (psi)	Rupture Time (hrs)	Total Measured Elongation (in.)
6Al-2V-1Mo	800	80,000	201.6	0.300
		70,000	306.2[a]	0.0605
	1020	48,500	2.4	c
		30,000	33.0	0.595
		20,000	149.6[b]	0.325
6Al-2V-2Mo	800	80,000	276.3	c
		60,000	319.1[a]	0.0039
	1020	50,000	9.4	0.410
		40,000	19.8	0.470
		30,000	45.8	0.630
		20,000	248.6	0.800
6Al-3V-1Mo	800	80,000	157.6	0.295
		70,000	304.8[a]	0.0462
		60,000	345.8[a]	0.0231
	1020	50,000	5.6	0.500
		40,000	4.9(?)	0.375
		30,000	27.0	0.820
		20,000	123.0	1.110

* Alloys heat treated: 1470°F-6 hrs-AC, 1020°F-24 hrs-AC.

a Test stopped at time indicated. No failure.

b Test stopped due to excessive elongation.

c Failure occurred outside gage marks.

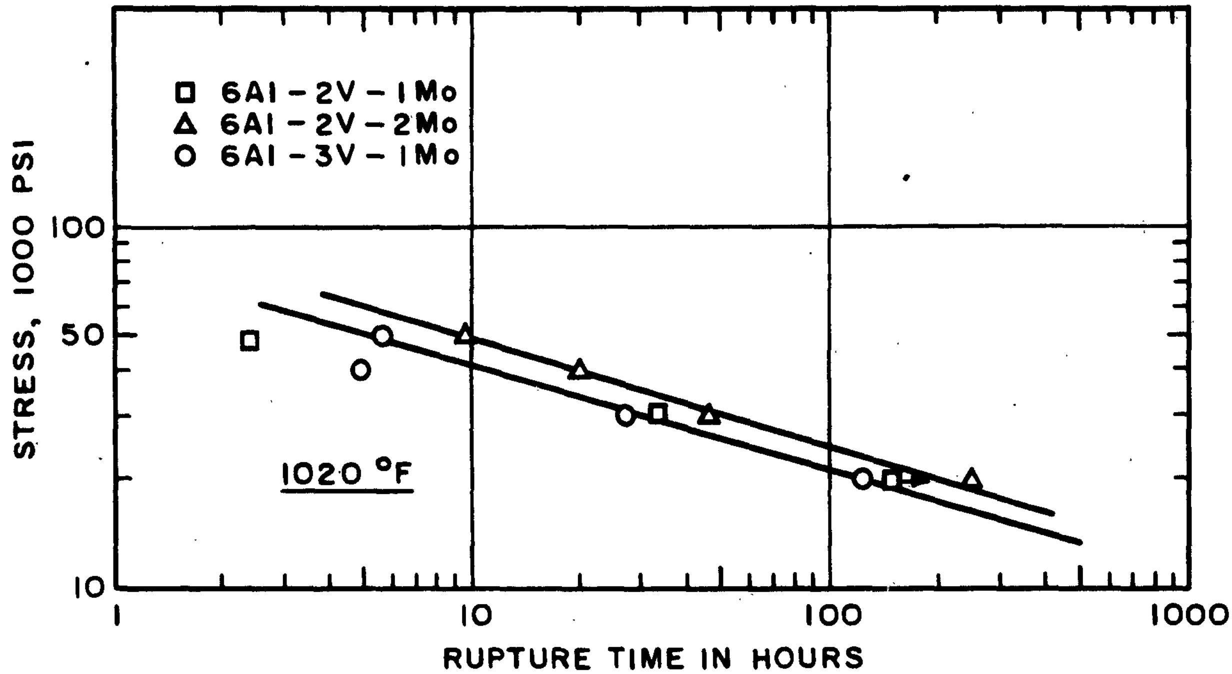


FIG. 33— STRESS-RUPTURE CURVES FOR ALLOYS Ti-6%Al-2%V-1% AND 2%Mo AND Ti-6%Al-3%V-1%Mo AT 1020 °F

800°F-60,000 psi-319.1 hours had essentially the same ultimate strength (153,000 psi) as the unexposed alloy specimen and comparable ductility (12% exposed versus 11% unexposed). Both exposed specimens of the 6Al-3V-1Mo alloy failed in imperfect areas outside the gage marks. The ultimates were around 157,000 psi. Because of the position of the fractures the ductilities are not as high as normally expected. The specimen creep tested at a stress of 70,000 psi showed 3% elongation between gage marks, while the specimen exposed under a stress of 60,000 psi had 7% elongation between gage marks. Based on the latter value it would appear that this alloy and the other modifications are reasonably stable materials.

4. Bend Properties

Guided bend tests consisting of bending 0.060 in. thick sheet specimens around mandrels of varying radius were conducted on the alloys in the heat treated condition: 1470°F-6 hrs-AC. The 6Al-2V-1Mo and 6Al-2V-2Mo alloys passed 3T; each failed trials at 2T at an angle of 70°. The 6Al-3V-1Mo passed 2T.

D. Properties of Ti-7Al-3Mo Alloy

Under Contract No. AF 33(038)-22806 a forging study was made of the Ti-7Al-3Mo alloy (see WADC TR 54-278 Part 2).(4) Wide variations in tensile ductility as expressed by reduction in area were obtained with differing thermal and working history. In an effort to isolate the variables influencing ductility in this alloy, samples representing each of the thermal and working histories were analyzed for hydrogen. In addition, nitrogen and oxygen determinations were made on a series of samples representing a study of the effect of the amount of reduction at the initial and finishing temperatures of forging. The hydrogen analyses were made by the Materials Laboratory, Wright Air Development Center. The hydrogen contents are plotted versus tensile reduction in area in Figure 34. There appears to be some correlation between hydrogen content and ductility. However, the wide range in ductility for a given hydrogen level suggests that there are other factors operative. Both equiaxed and Widmanstätten structures were observed. This gross structural difference does not appear to be of significance.

The results of the hydrogen, nitrogen and oxygen analyses on samples from the study of working variables is given in Table 16. No single element gives a correlation with ductility for both the odd numbered group of specimens (finished at 1800°F) and the even numbered group (finished at 1650°F); nor does the total atomic percentage of the elements H, N and O provide a correlation with ductility.

In a forging study of the 7Al-3Mo alloy referred to above, it was found that in the two cases in which it was applied, the heat treatment: 1560°F-4 hours-air cool, 1020°F-24 hours-air cool, resulted in significantly better ductility than the heat treatment: 1610°F-24 hours-air cool, 1020°F-24 hours-air cool. For these results refer to Table 25 of WADC TR 54-278 Pt 2. The results of another part of this forging study are summarized in Figure 35 (from WADC TR 54-278 Pt 2). The objective of this part was to determine the effects of finishing temperature and amount of reduction at the finishing temperature on tensile ductility. Finishing temperatures were 1800° and 1650°F. Amounts of

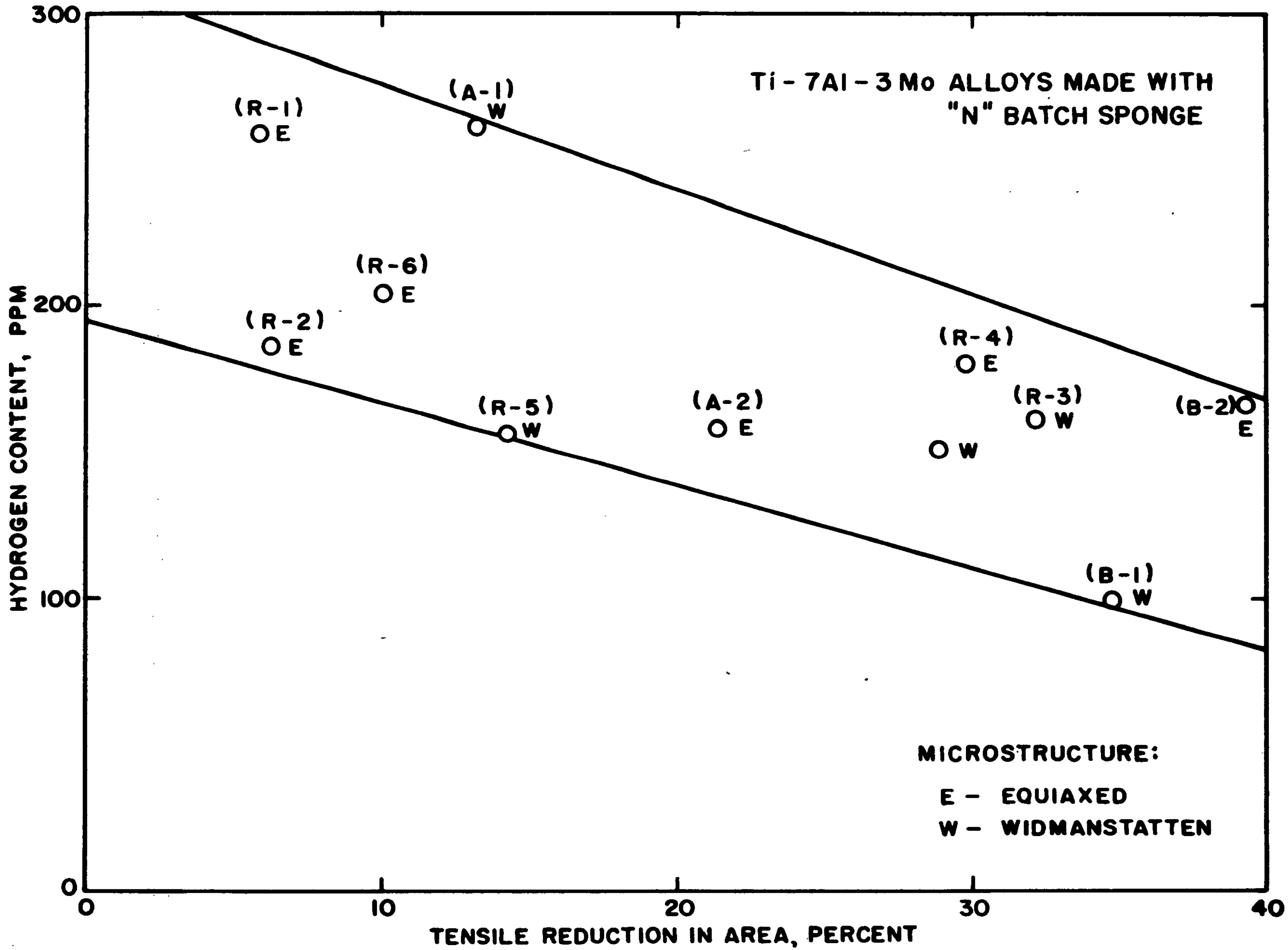


FIG. 34-CORRELATION OF HYDROGEN CONTENT WITH TENSILE DUCTILITY FOR Ti - 7% Al - 3% Mo (4N) ALLOY

TABLE 16

CONTENT OF INTERSTITIALLY SOLUBLE ELEMENTS, H, N AND O

IN SPECIMENS FROM FORGING STUDY OF Ti-7Al-3Mo ALLOY

Sample	Working History	Tensile Reduction in Area %	H		N		O		Combined
			wt. %	at. %	wt. %	at. %	wt. %	at. %	at. %
R-1	3" to 1" at 2000°F, 1" to 1/2" at 1800°F*	5.9+	0.0259	1.22	0.042	0.144	0.192	0.58	1.94
R-3	3" to 3/4" at 2000°F, 3/4" to 1/2" at 1800°F	32.1	0.0161	0.76	0.019	0.065	0.060	0.18	1.00
R-5	3" to 5/8" at 2000°F, 5/8" to 1/2" at 1800°F	14.2	0.0156	0.74	0.019	0.065	0.030	0.090	0.89
R-2	3" to 1" at 2000°F, 1" to 1/2" at 1650°F	6.3	0.0187	0.89	0.026	0.089	0.040	0.120	1.10
R-4	3" to 3/4" at 2000°F, 3/4" to 1/2" at 1650°F	29.8	0.0181	0.86	0.017	0.058	0.038	0.114	1.03
R-6	3" to 5/8" at 2000°F, 5/8" to 1/2" at 1650°F	10.1	0.0204	0.97	0.027	0.092	0.080	0.238	1.30

* All dimensions are diameters of rounds.

\+ All values are average for two tests except value for R-2 specimen.

All specimens heat treated 1610°F-24 hours-air cool, 1020°F-24 hours-air cool before testing.

reduction were 37, 55.5 and 75%. Tensile specimens representing the various conditions were given the heat treatment: 1610°F-24 hours-air cool, 1020°F-24 hours-air cool before testing. Figure 35 indicates 55.5% reduction to be optimum. The other reductions, 37 and 75%, resulted in poor ductilities. The purpose of the present work was to evaluate the effect of the heat treatment: 1560°F-4 hours-air cool, 1020°F-24 hours-air cool on these materials. The results are presented in Table 17.

Only three of the original six conditions are represented. There was not sufficient stock remaining for the other three conditions to make one or more tensile specimens. Fortunately, the two conditions resulting in the poorest ductilities (about 6% tensile reduction in area) are among those represented. These conditions resulted from reducing 75% at the finishing temperatures of 1800° and 1650°F, i.e., forging procedures R-1 and R-2, respectively. The third condition represented resulted in the best ductility (32% tensile RA) in the original work. All specimens given the 1560°F for 4 hours coarsening anneal had more than 30% tensile RA. One specimen each for the three conditions was vacuum annealed at 1560°F for 4 hours to remove hydrogen simultaneously with the coarsening anneal. The average ductility, as measured by reduction in area for the three specimens, showed a 10% relative increase over the average of the four specimens given the straight 1560°F-4 hours anneal followed by aging. For the thermal histories represented in the forging study there appears to be little doubt that compared to the 1610°F-24 hours anneal, the 1560°F-4 hours anneal results in considerably better and more uniform tensile ductilities.

However, these results are at variance with results of a heat treatment study made on the 7Al-3Mo alloy, see Table 20, WADC TR 54-278 Pt 2. The material used in this study was not the same heat as was used for the forging study. The ingot was forged from 3 in. round to 1 in. square at 2000°F and finished at 1950°F to 1/2 in. round. To check the previous results, some more specimens of this material were given the heat treatments under discussion. The results given in Table 18 confirm the previous findings that the 1610°F for 24 hours coarsening anneal produced better ductility than the 1560°F for 4 hours anneal. Also, two specimens initially annealed at 1610°F for 24 hours, aged and then annealed at 1560°F for 4 hours and aged, did not show improved ductility over specimens given one or the other of the double heat treatments. The purpose of the double heat treatment was to superimpose the 1560°F α/β ratio on the structure obtained by the 1610°F for 24 hours coarsening anneal. The higher finishing temperature of this material compared to those used in the forging study resulted in significantly different microstructures, as may be seen by comparing Figures 36, 37 and 38. Structures produced by the 1610°F anneal, 1020°F age were similar but more coarse. The material used in the heat treatment study was finished in the β, while the forging study material was finished at two temperatures in the α-β field. This may be the significant factor in the difference in heat treatment response observed. It may be noted that microstructures resulting from the forging study have a considerably greater average mean free path in the α phase than that of the heat treatment study. Also, the latter structure has a relatively large prior β grain size, while the prior β structures of the former microstructures have been completely obliterated.

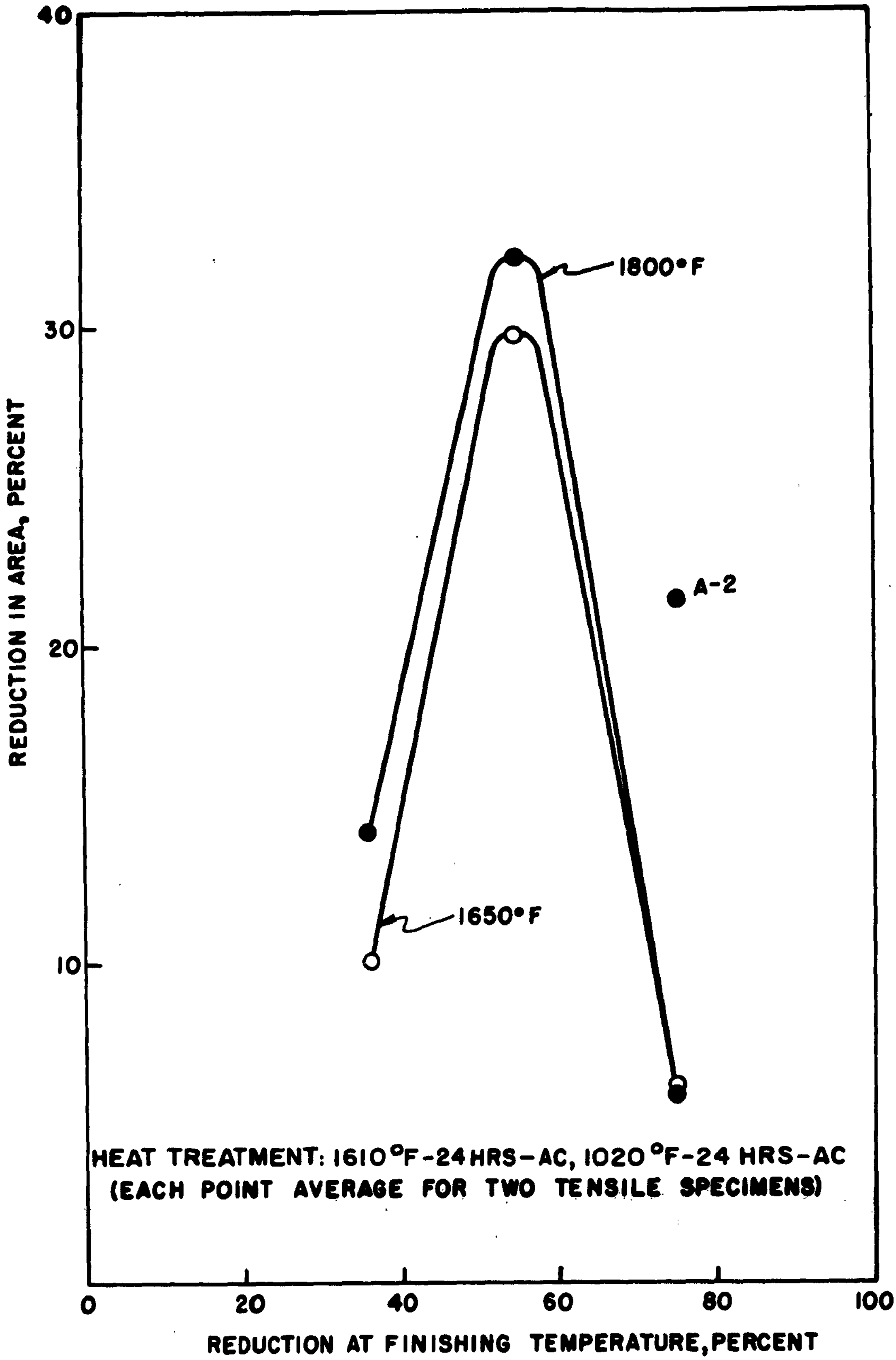


FIG. 35-INFLUENCE OF FINISHING TEMPERATURE AND PER CENT REDUCTION IN AREA ON THE TENSILE REDUCTION IN AREA OF Ti-7%Al-3%Mo (4N) ALLOY.

TABLE 17

HEAT TREATMENT COMPARISON FOR Ti-7Al-3Mo (4N) MATERIAL

Forging Procedure	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Fracture Stress psi
		1610°F-24 hrs-AC, 1020°F-24 hrs-AC*			
R-1	172,000	158,000	6.3	3.0	184,000
	176,000	160,000	5.5	3.0	187,000
R-2	156,000	146,000	6.3	2.0	165,000
R-3	160,000	136,000	34.9	15.0	216,000
	163,000	136,000	29.3	14.5	208,000
		1560°F-4 hrs-AC, 1020°F-24 hrs-AC			
R-1	155,000	137,000	44.4	18.0	216,000
R-2	158,000	146,000	31.9	13.5	204,000
R-3	159,000	141,000	32.7	15.0	208,000
	160,000	142,000	38.1	16.0	213,000
		VA 1560°F-4 hrs-AC, 1020°F-24 hrs-AC			
R-1	155,000	137,000	38.8	15.0	210,000
R-2	158,000	146,000	43.2	19.0	214,000
R-3	154,000	138,000	41.1	16.0	205,000

R-1 Forged from 3 to 1 in. round (89% reduction) at 2000°F, and finished to 1/2 in. round (75% reduction) at 1800°F.

R-2 Forged from 3 to 1 in. round (89% reduction) at 2000°F, and finished to 1/2 in. round (75% reduction) at 1650°F.

R-3 Forged from 3 to 3/4 in. round (94% reduction) at 2000°F, and finished to 1/2 in. round (55.5% reduction) at 1800°F.

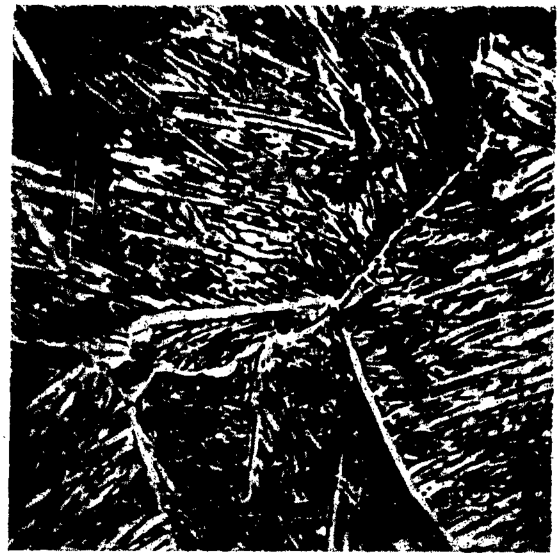
* Data for this heat treatment from WADC TR 54-278 Pt 2 (AF 33(038)-22806).

TABLE 18

HEAT TREATMENT COMPARISON FOR Ti-7Al-3Mo (1N) MATERIAL

Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elongation %	Fracture Stress psi
		1610°F-24 hrs-AC, 1020°F-48 hrs-AC		
163,000	144,000	28.0	14.0	210,000*
163,000	143,000	25.8	15.5	206,000
		1560°F-4 hrs-AC, 1020°F-48 hrs-AC		
160,000	134,000	21.6	14.0	194,000*
162,000	142,000	16.0	13.0	190,000
164,000	144,000	23.8	13.0	204,000
		1610°F-24 hrs-AC, 1020°F-48 hrs-AC, 1560°F-4 hrs-AC, 1020°F-48 hrs-AC		
164,000	143,000	10.8	8.0	184,000
162,000	144,000	21.6	15.0	199,000

* These data from Table 20. WADC TR 54-278 Pt 2. Specimen aged for 24 hours at 1020°F. This is not believed to result in real difference in properties compared to aging for 48 hours.



Neg. No. 11212 X 75(

Fig. 36

Ti-7Al-3Mo (1N), 80% reduction by forging at finishing temperature of 1950°F. Heat treatment: 1560°F-4 hrs-AC, 1020°F-48 hrs-AC. α + β.

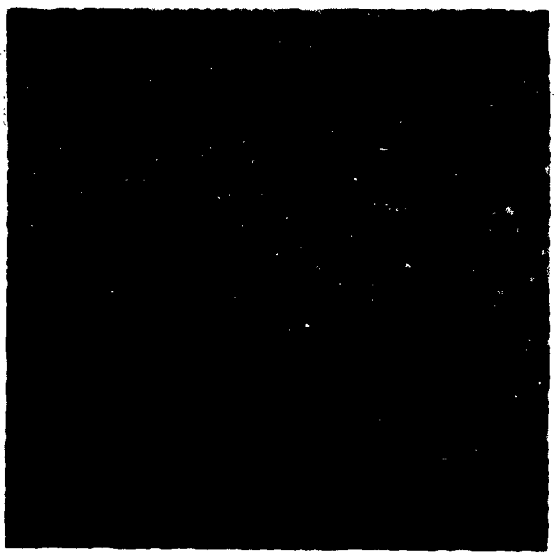


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Fig. 37

Ti-7Al-3Mo (4N), 75% reduction by forging at finishing temperature of 1800°F. Heat treatment: VA 1560°F-4 hrs-AC, 1020°F-24 hrs-AC. α + β.

Neg. No. 11214 X 75(

Fig. 38

Ti-7Al-3Mo (4N), 75% reduction by forging at finishing temperature of 1650°F. Heat treatment: VA 1560°F-4 hrs-AC, 1020°F-24 hrs-AC. α + β.

Etchant: 60 cc glycerine, 20 cc HNO_3, 20 cc HF.

E. Factors Affecting Ductility of Titanium-Aluminum Alloys in the Range from 6 to 10% Aluminum

1. Introduction

In previous work under Contract No. AF 33(616)-2351(5) the tensile properties, particularly ductility, of Ti-8Al were found to be very dependent upon the temperature from which the specimens were quenched. The relationships found for reduction in area and proportional limit as functions of annealing temperature are shown in Figure 39. In this work annealing time was increased as annealing temperature was lowered in order to compensate somewhat for the decreasing diffusion rate. With decreasing annealing temperature tensile reduction in area decreased from about 35% to zero for annealing temperatures in the range from 1470° to 1020°F, and increased to the 32-34% level in the range from 1020° to 750°F where it was constant down to 390°F. Vacuum annealing to remove hydrogen did not appear to lessen susceptibility to embrittlement when the alloy was annealed at the critical temperature of 1020°F.

Upon annealing resistivity specimens at successively lower temperatures and measuring resistivity at room temperature after each anneal, the relationship shown in Figure 40 was obtained. Annealing times were the same as used in the tensile study summarized in Figure 39. The drop in resistivity corresponds to the loss in tensile ductility and indicates that some change takes place in the alloy solid solution.

2. Heat Treatments, Room Temperature Properties and Microstructures

During the present work the above mentioned resistivity specimens were reheated to 1470°F, quenched and annealed at 750°F for 48 hours. The measured resistivities are included in Figure 40. The resistivities were at the level of the specimens as quenched from 1470°F. This means that the state of the alloy is not changed by annealing at 750°F for 48 hours following a quench from 1470°F. Thus, the apparent recovery in ductility by annealing at 750°F or lower is due simply to the fact that the annealing times were not sufficient for the embrittling reaction to occur.

The present work was primarily concerned with studying the effect of heat treatment on the mechanical properties and microstructures of 6 and 10Al alloys and with further study of the 8Al composition as an iodide titanium-base material as well as a sponge base material.

Tensile properties of the 6 and 10Al alloys as quenched from various annealing temperatures are given in Tables 19 and 20, respectively. These tensile data were obtained generally at two loading rates, 600 and 4000 lbs/min, in order to get different magnitudes of strain rate. A few tests of the 10Al alloy were run at a loading rate of 1000 lbs/min.

The data for the 6Al alloy showed no significant variation in any tensile property with varying annealing temperature. Also, no significant difference in properties was noted for the different loading rates. Tensile reduction in area values plotted versus annealing temperatures are shown in Figure 41.

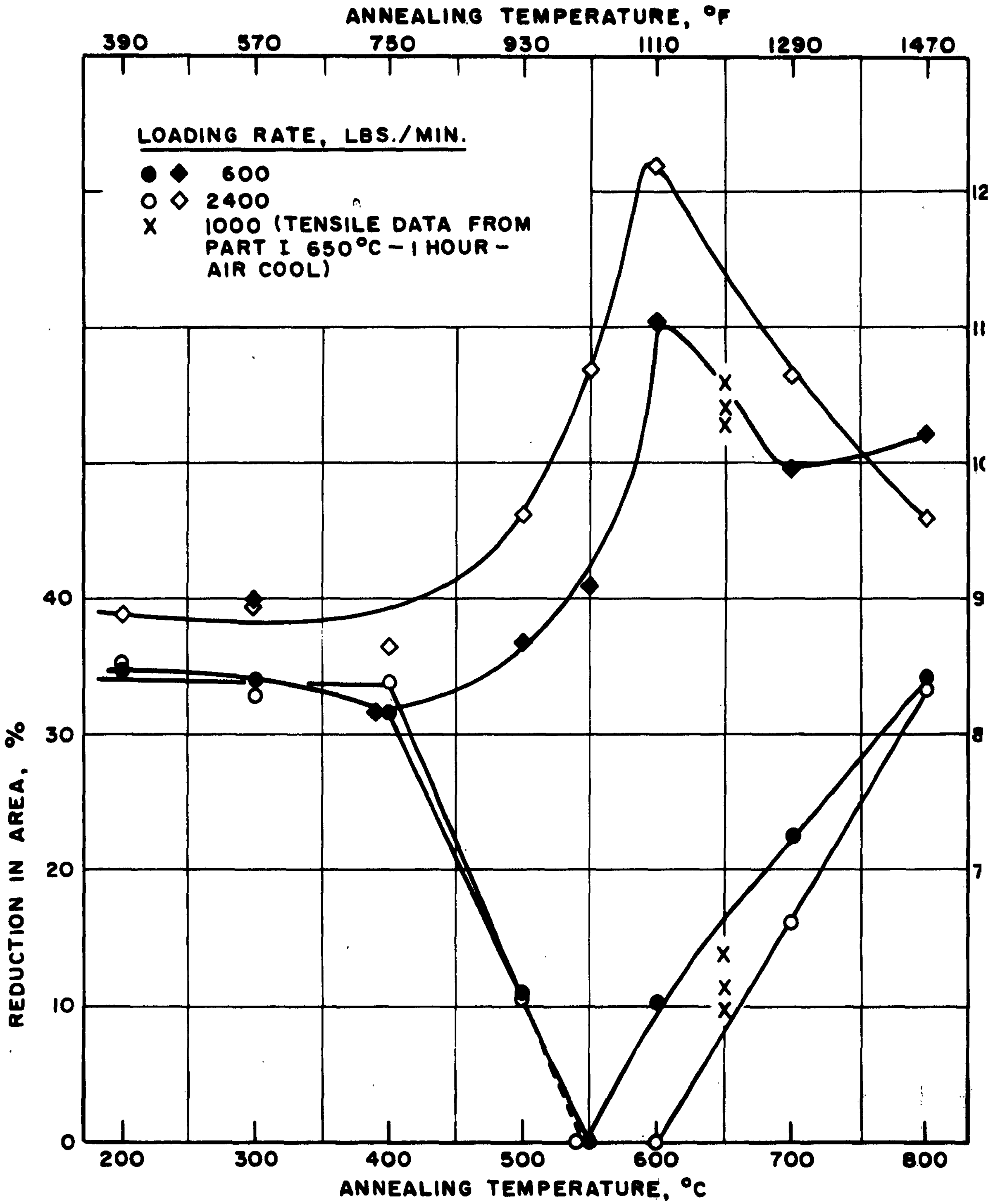


FIG. 39-PROPORTIONAL LIMIT AND REDUCTION IN AREA VS. ANNEALING TEMPERATURE FOR Ti-8%Al (2N) ALLOY.

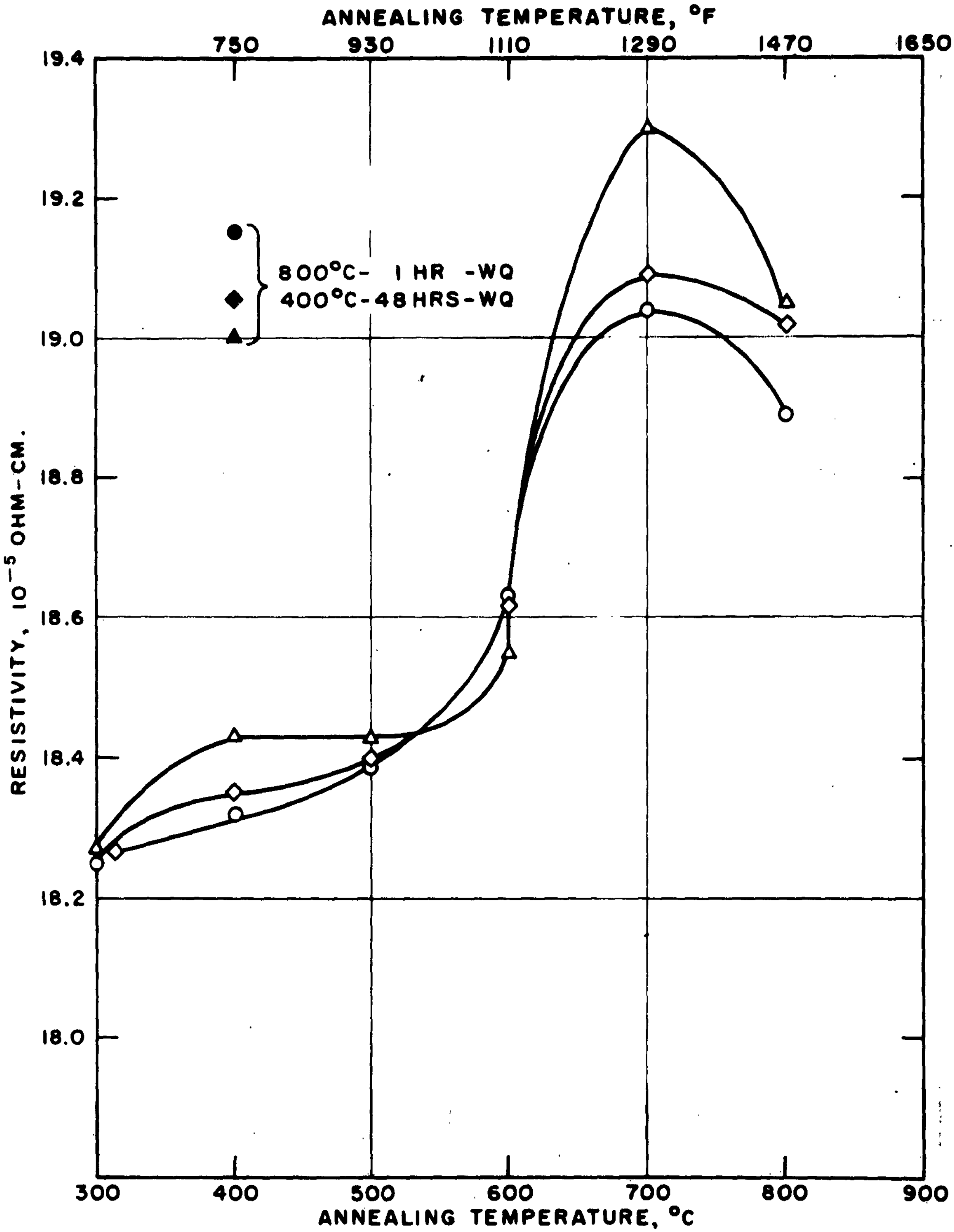


FIG. 40 - RESISTIVITY VS. ANNEALING TEMPERATURE FOR Ti-8% Al (2N) ALLOY. SPECIMENS ANNEALED AT SUCCESSIVELY LOWER TEMPERATURES.

TABLE 19

EFFECT OF ANNEALING TEMPERATURE

ON THE TENSILE PROPERTIES OF Ti-6Al (1BB) ALLOY

Heat Treatment	Loading Rate lbs/min	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Propor-tional Limit	Reduction in Area %	Elong. %	Modulus of Elasticity (psi x 10^{-6})	Fracture Stress psi
1470°F-1 hr-WQ	600	133,000	113,000	88,000	36.0	17.0	17	174,000
		139,000	117,000	92,000	34.0	17.0	18	180,000
	4000	136,000	122,000	96,000	30.1	17.0	17	170,000
		145,000	134,000	106,000	43.0	17.0	20	206,000
1290°F-1 hr-WQ	600	137,000	116,000	92,000	33.2	18.0	18	174,000
		139,000	116,000	94,000	33.1	17.0	17	179,000
	4000	139,000	122,000	96,000	30.5	18.0	18	169,000
		143,000	132,000	100,000	42.0	17.0	18	192,000
1110°F-2 hrs-WQ	600	137,000	115,000	88,000	26.6	18.0	20	180,000
		143,000	121,000	102,000	34.8	16.0	23	182,000
	4000	143,000	126,000	104,000	28.0	14.0	18	190,000
		143,000	131,000	108,000	27.0	16.5	19	191,000
1020°F-16 hrs-WQ	600	139,000	122,000	107,000	31.5	15.0	16	178,000
		140,000	119,000	90,000	34.5	17.0	17	181,000
	4000	142,000	130,000	106,000	35.0	14.5	18	184,000
		142,000	130,000	109,000	30.5	13.0	17	180,000
750°F-48 hrs-WQ	600	146,000	121,000	97,000	35.7	15.0	18	189,000
		140,000	114,000	75,000	33.6	15.0	17	185,000
	4000	144,000	128,000	104,000	31.0	14.0	15	187,000
		143,000	125,000	---	29.3	15.0	18	183,000

TABLE 20

EFFECT OF ANNEALING TEMPERATURE
ON THE TENSILE PROPERTIES OF Ti-10Al (1BB) ALLOY

Heat Treatment	Loading Rate lbs/min	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Propor-tional Limit	Reduction in Area %	Elong. %	Modulus of Elasticity (psi x 10^{-6})	Fracture Stress psi
2010°F-1/2 hr-IBQ	1000	134,000	128,000	73,000	a	0.4	19	133,000
1830°F-1 hr-IBQ	1000	121,000	113,000	80,000	7.5	2.0	18	127,000
1830°F-1 hr-WQ	600	130,000	125,000	95.000	---	0.3	16	130,000
	1000[b]	127,000	114,000	88,000	3.2	1.0	17	132,000
	4000	105,000	105,000[c]	---	---	0.14	19	105,000
1650°F-1 hr-WQ	600	124,000	122,000	93,000	---	0.3	18	124,000
	4000	126,000	122,000	93,000	9.3	2.5	20	139,000
1470°F-1 hr-WQ	600	67,000	---	---	0.0	0.0	16	67,000
		106,000	---	106,000	---	0.08	19	106,000
	4000	80,000	---	---	0.0	0.0	18	80,000
1290°F-1 hr-WQ	600	77,000	---	---	0.0	0.0	19	77,000
		68,000	---	---	0.0	0.0	18	68,000
1110°F-2 hrs-WQ	600	54,000	---	---	0.0	0.0	21	54,000
		72,000	---	---	0.0	0.0	18	72,000
	4000	96,000	---	---	0.0	0.0	20	96,000
		63,000	---	---	0.0	0.0	21	63,000

a - Fractured in threads.

b - Tested within an hour of quench.

c - 0.1% offset yield strength.

IBQ - Ice Brine Quench.

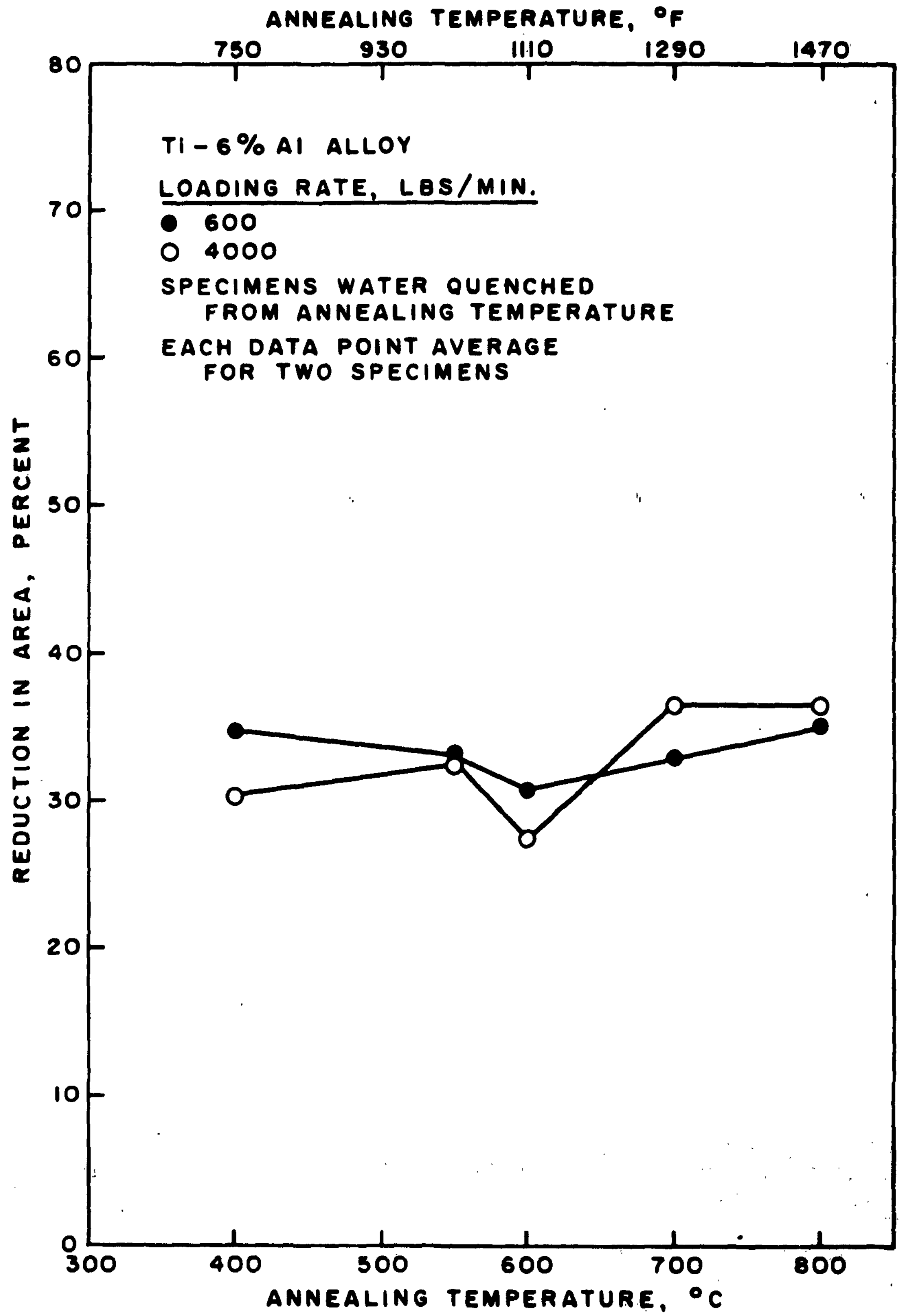


FIG. 41 - REDUCTION IN AREA VS. ANNEALING TEMPERATURE FOR Ti - 6% Al (IBB) ALLOY

Most of the 10Al specimens were brittle. Yield strength values were surprisingly low. The average of the 0.2% offset yield values was 120,000 psi compared to 131,000 psi obtained for 8Al specimens. Fracture stress generally increased with increasing annealing temperature as illustrated in Figure 42. A few specimens, quenched from 1650°F or higher, exhibited measurable, but small, ductilities, indicating perhaps that the alloy is not inherently brittle.

Two 8Al (sponge) tensile specimens were given the 1020°F-48 hours embrittling heat treatment and then annealed at 1470°F for 2 hours and water quenched. Tensile properties were as follows:

Specimen No.	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elong. %	Fracture Stress psi
(1)	144,000	126,000	33.8	13.0	191,000
(2)	146,000	129,500	29.2	13.0	184,000

Comparison of these properties with those for 8Al specimens given only the 1470°F-1 hr-WQ heat treatment(5) show that tensile properties including ductility are completely recovered.

Tensile test results for 8 and 10Al specimens vacuum annealed to remove hydrogen are summarized in Table 21. Results of hydrogen and oxygen analyses are included. The hydrogen contents of these specimens were generally around 20 ppm, the lowest being 12 ppm. A Ti-8Al (2N) specimen heat treated to an initially ductile condition (vacuum annealed at 1470°F for 48 hours-water quenched, aged at 750°F for 64 hours-air cooled) was aged at 750°F under a stress of 30,000 psi for 1006 hours. The stress-aging treatment caused embrittlement. The microstructure of this specimen shown in Figure 43 indicates copious precipitate.

A group of iodide titanium-base 8Al tensile specimens were vacuum annealed at 1560°F for 6 hours and water quenched. One was tested as-quenched, one as-aged at 1020°F for 42 hours, and a third as-aged at 1020°F for 500 hours. The as-quenched specimen was highly ductile and no precipitate was apparent in its microstructure. The specimen aged for 42 hours had a lower yield strength and much poorer ductility than the as-quenched specimen. The microstructure at 250 X was indicated to contain small particles in the grain boundaries (see Figure 44). Applying a dynamic vacuum during aging at 1020°F for 48 hours also resulted in very low ductility. The specimen aged for 500 hours had a still lower yield strength and was brittle. Its microstructure at 250 X was the same as that of the specimen aged for 42 hours. However, better appreciation of the extent of precipitation and the size of the particles is realized at 1500 X (see Figure 45). Another specimen of this material vacuum annealed at 1650°F for 96 hours-air cooled and aged at 1020°F for 48 hours had lower yield strength but slightly improved ductility compared to the specimen just discussed. The microstructure of this specimen is shown in Figure 46. Only about one-half of the periphery of the fracture showed signs of plastic deformation.

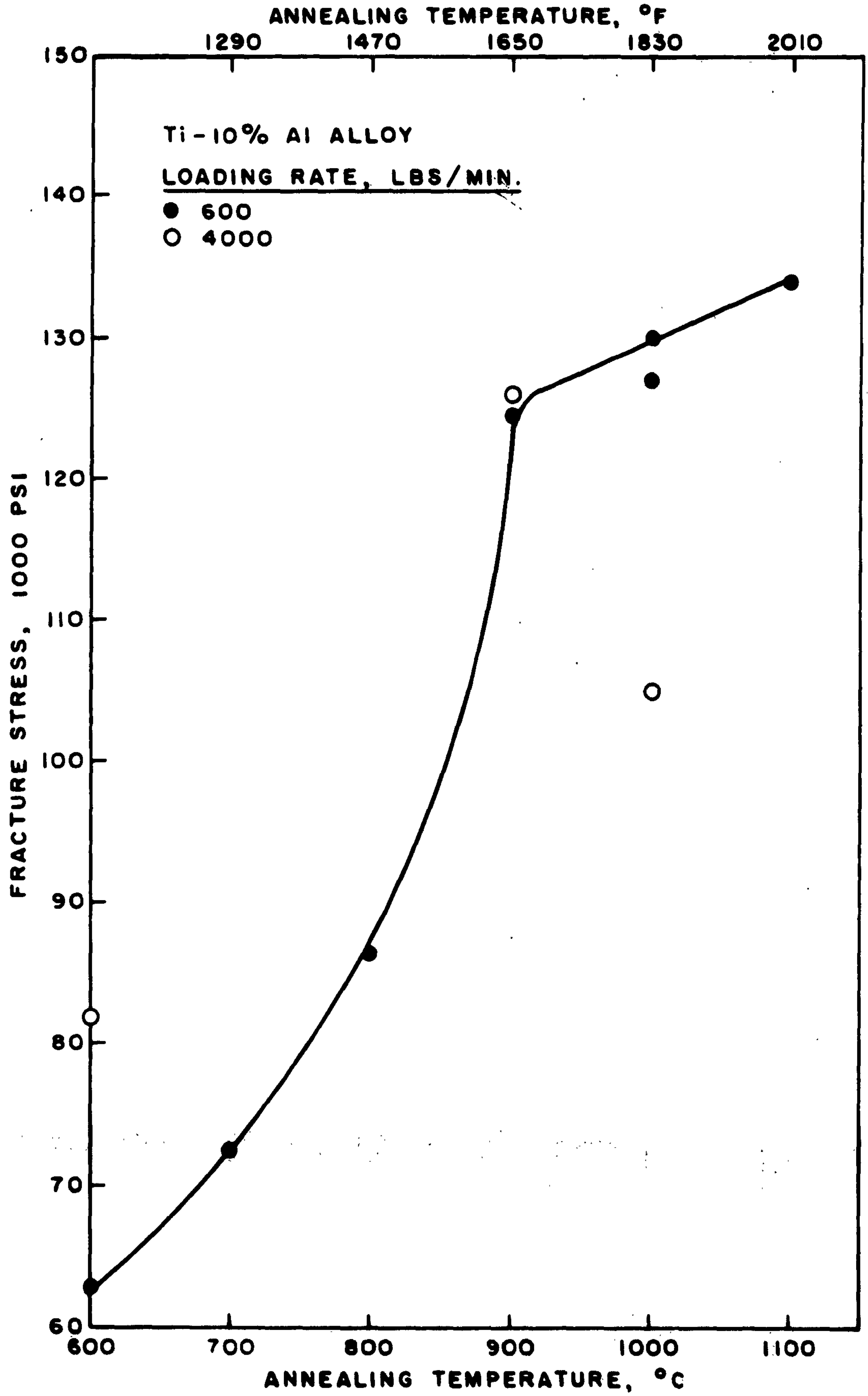


FIG. 42 – FRACTURE STRESS VS. ANNEALING TEMPERATURE FOR Ti-10% Al (IBB) ALLOY

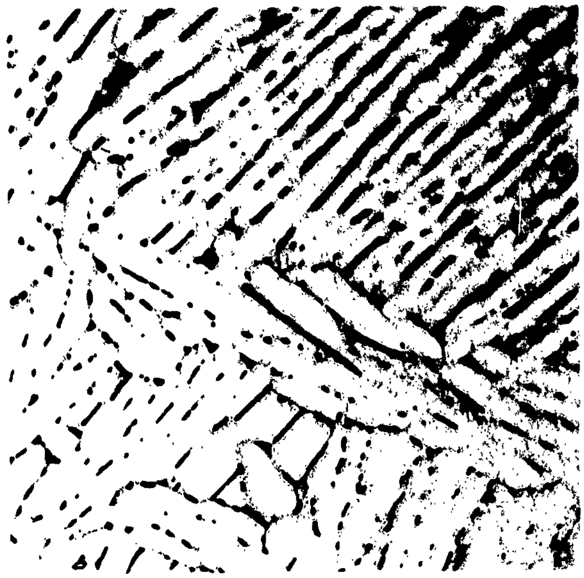
TABLE 21

EFFECT OF VACUUM ANNEALING ON TENSILE PROPERTIES

OF Ti-8Al AND Ti-10Al ALLOYS

Alloy	Heat Treatment	Ultimate Tensile Strength psi	Yield Strength (0.2% Offset) psi	Reduction in Area %	Elong. %	Fracture Stress psi	Hydrogen Content ppm	Oxygen Content wt. %
Ti-8Al (2N)	VA 1470°F-48 hrs-WQ, 750°F-64 hrs-AC Stress aged: 750°F-30,000 psi-1006 hrs	84,000	----	0.0	0.07	84,000	--	---
Ti-8Al (Iod)	VA 1560°F-6 hrs-WQ	116,000	99,000	31.7	16.0	148,000	--	---
	VA 1560°F-6 hrs-WQ, 1020°F-42 hrs-WQ	91,000	88,000	10.2	2.0	80,000	37	---
	VA 1560°F-6 hrs-WQ, 1020°F-500 hrs-WQ	62,000	59,000[a]	--	0.14	62,000	21	0.031
	VA 1650°F-96 hrs-AC, 1020°F-48 hrs-AC	53,000	48,000	3.2	1.0	55,000	18	0.060
	VA 1560°F-6 hrs-WQ, VA 1020°F-48 hrs-WQ	112,000	108,000	2.2	1.0	113,000	--	---
Ti-8Al (2N)	VA 2550°F-3/4 hr-FC, 1020°F-48 hrs-AC	88,000 64,000	86,000 58,000	4.8 5.5	0.5 1.5	93,000 17,000	23 --	0.097 ---
Ti-10Al (1BB)	VA 2550°F-3/4 hr-FC, 1020°F-48 hrs-AC	24,000 59,000	---- ----	0.0 0.0	0.0 0.0	24,000 59,000	24 --	--- ---
	VA 1650°F-96 hrs-AC, 1020°F-48 hrs-AC	36,000 36,000	---- ----	0.0 0.0	0.0 0.0	36,000 36,000	12 --	0.048 0.049

a 0.1% offset yield.



Neg. No. 11521 X 1000

Fig. 43

Ti-8Al (2N), VA 1470°F-48 hrs-WQ, 750°F-64 hrs-AC. Stress aged: 750°F-30,000 psi-1006 hrs. α + precipitate.

Neg. No. 11220 X 250

Fig. 44

Ti-8Al (Iod), VA 1560°F-6 hrs-WQ, 1020°F-42 hrs-AC. α + precipitate.

Neg. No. 11252 X 1500

Fig. 45

Ti-8Al (Iod), VA 1560°F-6 hrs-WQ, 1020°F-500 hrs-AC. α + precipitate.

Neg. No. 11334 X 250

Fig. 46

Ti-8Al (Iod), VA 1650°F-96 hrs-AC, 1020°F-48 hrs-AC. α + precipitate.

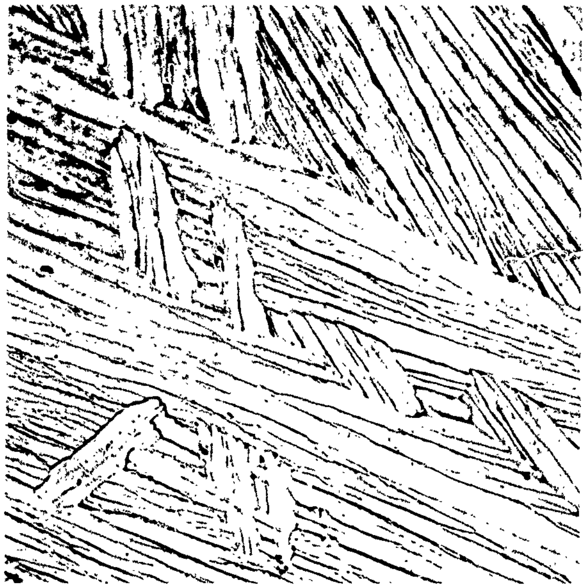
Etchant: 60 cc glycerine, 20 cc HNO_3, 20 cc HF.

In an effort to remove as much hydrogen as possible, two bars each of Ti-8Al and Ti-10Al were taken through essentially the same cycle in a vacuum fusion apparatus as was used for analyzing for hydrogen. The bars were 1/2 in. round by 3 in. long. The specimens, done one at a time, were wrapped in molybdenum foil to prevent contact with the graphite container and packing. Whereas in a hydrogen determination the sample is held at 2550°F for 15 minutes, these specimens were held for 45 minutes. After hydrogen removal, tensile specimens were machined from the bars. The specimens were aged at 1020°F for 48 hours. Tensile test results for these specimens are included in Table 21. The 8Al specimens exhibited low ductility and low yield strength. A microstructure representative of these two specimens is shown in Figure 47. The specimen with 88,000 psi ultimate tensile strength had a fracture similar to that described for the Ti-8Al (iodide) specimen vacuum annealed at 1650°F. It also showed plastic deformation around half of the periphery of the fracture. The fracture of the second specimen was more unusual in that after yielding, the load dropped steadily. The stress at 0.5% offset was 25,000 psi and fracture occurred at 17,000 psi. These fracture characteristics suggest that the yielding obtained at relatively low stress values was the result of crack formation and propagation occurring before the true yield strength of the matrix is reached. The cracks propagate accompanied by plastic deformation until conditions for catastrophic failure obtain, resulting in brittle fracture of the remaining sound metal.

Specimens of Ti-10Al subjected to the hydrogen removal treatment at 2550°F or vacuum annealed at 1650°F for 96 hours and subsequently aged at 1020°F for 48 hours were all brittle. The microstructures of the specimens vacuum treated at 2550°F were indicated to have relatively large quantities of precipitate (see Figure 48). The microstructure of a specimen vacuum annealed at 1650°F was much cleaner but also appeared to contain a small amount of very fine precipitate (see Figures 49 and 50).

3. Effect of Test Temperature on Mechanical Properties

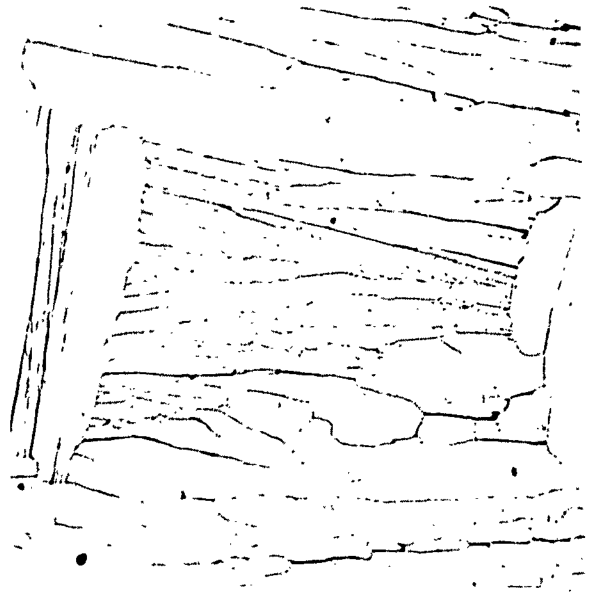
Tensile properties as a function of test temperature were studied to a limited extent for the 6, 8 and 10Al alloys. The heat treatment, 1020°F-48 hours-water quench, which severely embrittled the 8Al alloy, was applied to some of the specimens. In addition, the heat treatment, 1470°F-1 hour-water quench, which produced a ductile condition in the 8Al alloy, was applied to some 6Al alloy specimens. The tensile data are given in Table 22. Tensile reduction in area values are plotted versus test temperature in Figure 51. The 6Al alloy appeared to suffer a transition in ductility between 32° and 85°F for both heat treated conditions. The limited data for the 1020°F for 48 hours heat treated condition indicate that the transition range may be even more narrow, i.e., between 60° and 85°F. At temperatures below room temperature the 1470°F-1 hour-water quench condition was two to three times more ductile than the 1020°F-48 hours-water quench condition in agreement with results for 8Al alloy at room temperature. The 8Al alloy in the sensitized condition exhibited increasing ductility with increasing temperature. However, at 430°F the reduction in area was only 2/3 of the room temperature value in the 1470°F-1 hour-water quench condition. Specimens of the 10Al alloy in the sensitized condition were completely brittle at temperatures to 300°F. At 480°F a specimen exhibited 10% reduction in area.



Neg. No. 11219 X 250

Fig. 47

Ti-8Al (2N), VA 2550°F-45 min-FC, 1020°F-48 hrs-WQ. α + precipitate.



Neg. No. 11218 X 25

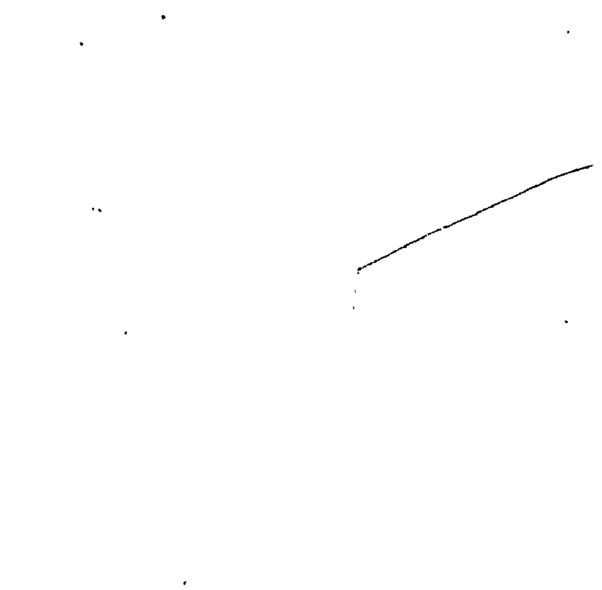
Fig. 48

Ti-10Al (1BB), VA 2550°F-45 min-FC, 1020°F-48 hrs-WQ. α + precipitate.

Neg. No. 11216 X 250

Fig. 49

Ti-10Al (1BB), VA 1650°F-96 hrs-AC, 1020°F-48 hrs-AC. α + barely observable precipitate in grain boundaries.



Neg. No. 11215 X 750

Fig. 50

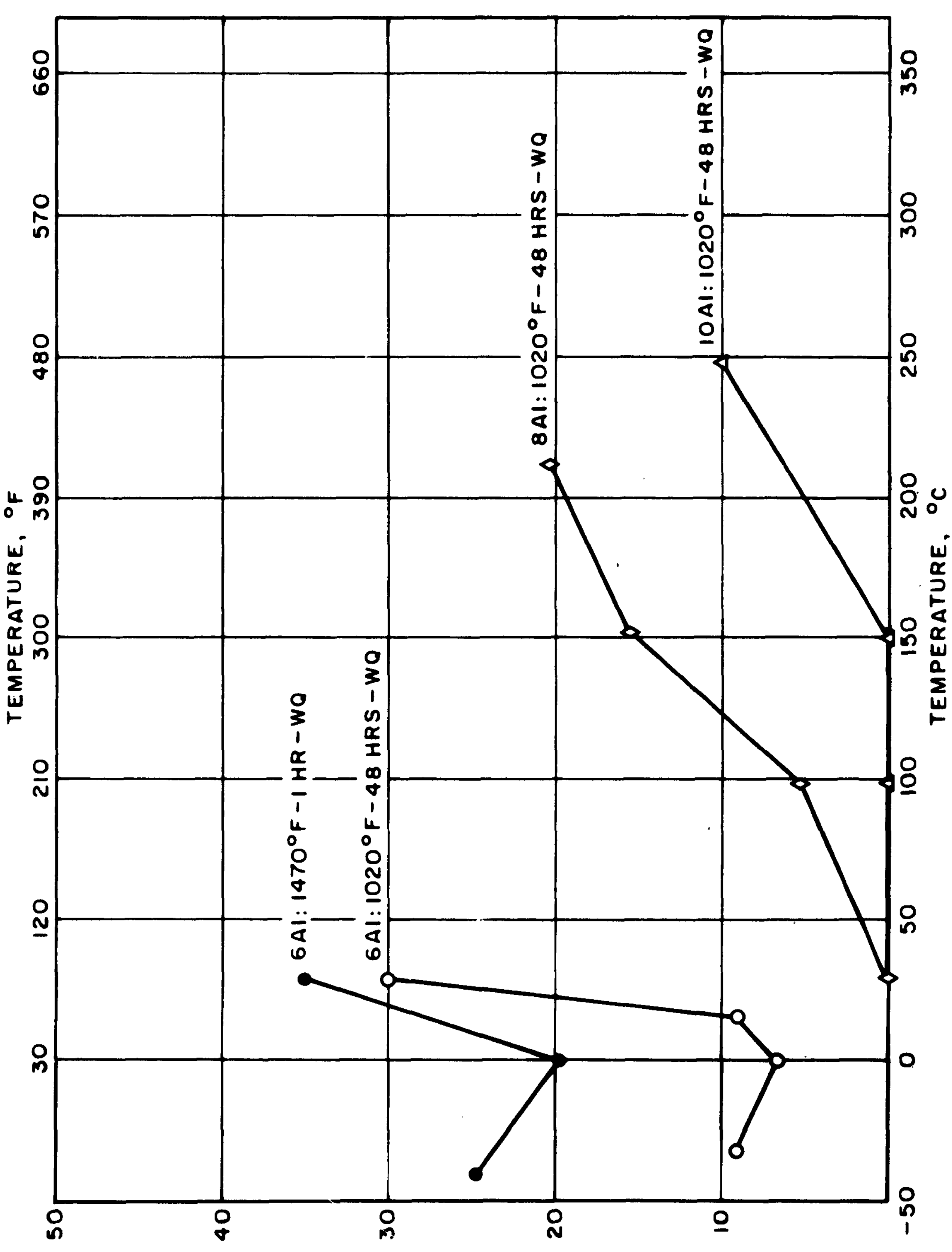
Ti-10Al (1BB), same as Fig. 49. The precipitate is a little more apparent at the higher magnification.

Etchant: 60 cc glycerine, 20 cc HNO_3, 20 cc HF.

TABLE 22

TEMPERATURE DEPENDENCE OF TENSILE PROPERTIES OF Ti-Al ALLOYS

Alloy	Heat Treatment	Test Temp (°F)	Ultimate Tensile Strength psi	Reduction in Area %	Elong. %	Fracture Stress psi
6Al (1BB)	1020°F-48 hrs-WQ	-25	158,000	9.0	3.5	174,000
		32	150,000	6.5	3.0	160,000
		60	151,000	9.0	4.0	166,000
		84	141,000	30.0	14.0	173,000
	1470°F-1 hr-WQ	-40	167,000	25.5	10.0	222,000
		32	142,000	19.5	13.0	171,000
		84	136,000	35.0	17.0	176,000
8Al (2N)	1020°F-16 hrs-WQ	84	110,000	0.0	0.0	110,000
	1020°F-48 hrs-WQ	210	126,000	5.5	3.0	133,000
		306	131,000	15.5	8.0	155,000
		396	112,000	20.0	10.5	142,000
10Al (1BB)	1020°F-48 hrs-WQ	210	73,000	0.0	0.0	73,000
		302	89,000	0.0	0.0	89,000
		478	113,000	10.0	3.0	125,000


TEMPERATURE, °F
30
120
210
300
390
480
570
660
TEMPERATURE, °C
-50
0
50
100
150
200
250
300
350
TENSILE REDUCTION IN AREA, PER CENT
50
40
30
20
10
6Al: 1470°F - 1 HR - WQ
6Al: 1020°F - 48 HRS - WQ
8Al: 1020°F - 48 HRS - WQ
10Al: 1020°F - 48 HRS - WQ

Notch impact as a function of temperature for the 6Al alloy is given in Figure 52. Impact values were more than twice as high for the 1470°F-1 hour-water quench condition as for the 1020°F-48 hours-water quench condition. Vacuum annealing to remove hydrogen resulted in no significant improvement of impact properties of specimens aged at 1020°F. The data are not sufficient to say whether a transition occurs in the temperature range 32° to 212°F covered by these tests.

The low temperature tensile and impact data for the 6Al alloy establish that it embrittles similarly to the 8Al alloy, but not as severely. The 10Al alloy is apparently embrittled in like fashion.

Some interesting microstructural details have been observed in this investigation. The discussion of these observations is included as an appendix, since they are more likely to obscure rather than elucidate the essential findings of this investigation.

4. Identification of the Embrittling Agent

Since long times in the temperature range 750° to 1020°F are apparently needed to develop conspicuous precipitate, a survey was made of the microstructures of creep-rupture specimens of 6Al (3B) and 8Al (1J) alloys. Specimens of the latter alloy tested at 800° and 1020°F showed indications of precipitate, with best indications given by specimens tested at 1020°F (see Figure 53). Three specimens of Ti-6Al (3B) were examined. These specimens had been tested at 800°F and had rupture lives of 148, 216 and 319 hours. No good indication of precipitation was found in the 148-hour rupture specimen; a trace of a precipitate was indicated in the 216-hour specimen; and rather good indication of a precipitate was found in the 319-hour specimen. However, the evidence of precipitation in this specimen was not as good as that shown in Figure 53.

In order to learn more about the precipitate specimens of 6, 8 and 10Al, alloys were vacuum annealed at 1560°F for 6 hours and air cooled. This treatment was followed by aging at 1020°F for 48 and 500 hours. Initial hydrogen contents were 233, 100 and 135 ppm for the 6, 8 and 10Al alloys, respectively. Another group of specimens was simultaneously heat treated except that the 1560°F anneal was carried out in evacuated Vycor bulbs. These specimens were examined metallographically. All were indicated to contain precipitate. The microstructures were similar to that shown in Figure 43. In general, the vacuum annealed specimens appeared to contain much less of the precipitate than the other specimens. Specimens of 16Al alloy were also examined after aging at 1020°F for 48 and 500 hours. These specimens had microstructures similar to those of the lower aluminum content alloys discussed above. However, the amount of precipitate in the 16Al alloy was indicated to be far less. This alloy analyzed 56 ppm hydrogen.

Several specimens in which the precipitate was highly developed were annealed at 1470°F for 2 or 3 hours and water quenched. The results are shown in Figures 54, 56 and 57. Figures 43, 45 and 55 show the corresponding structures before the resolutioning heat treatment. There is no difference in the structures.

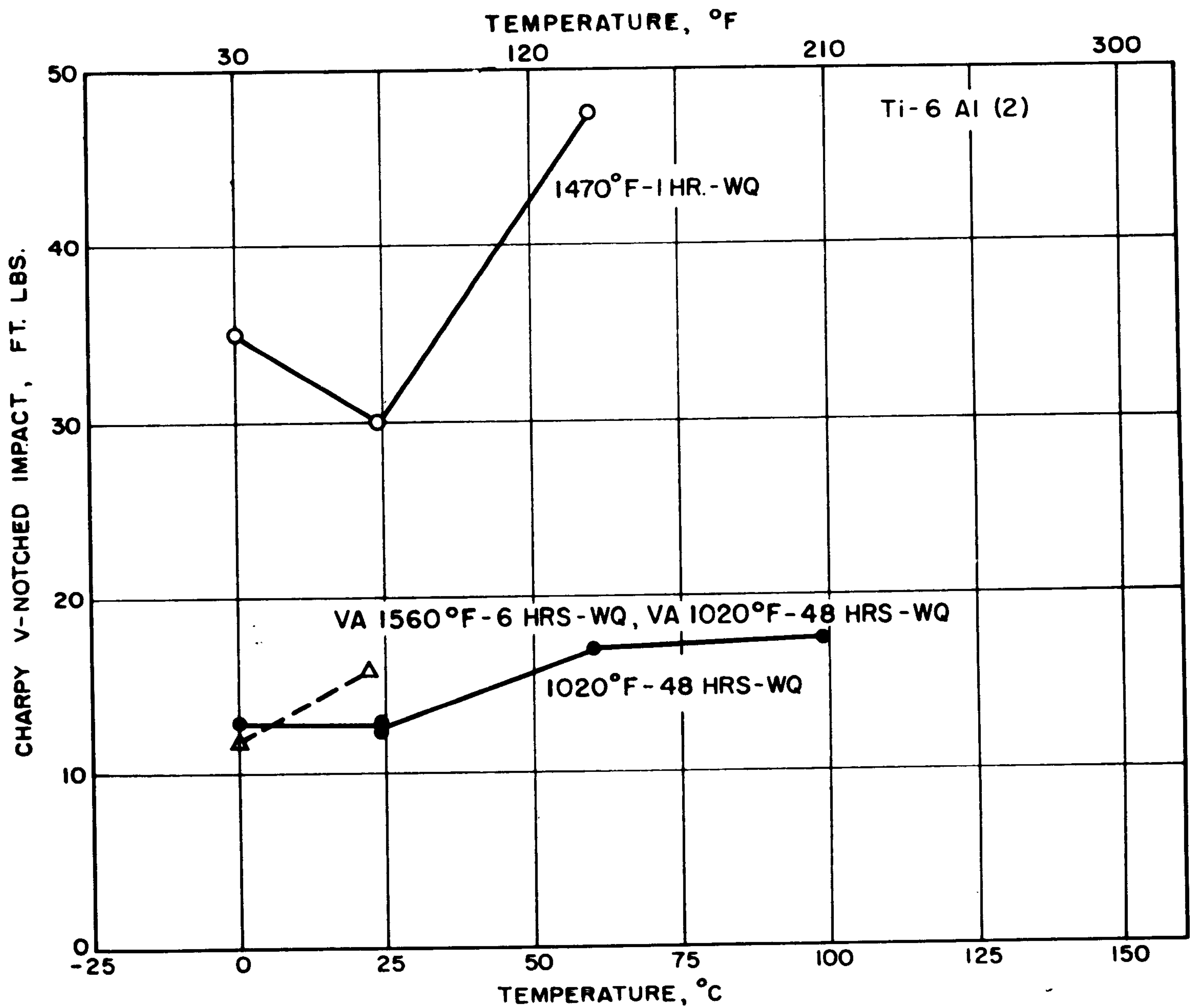
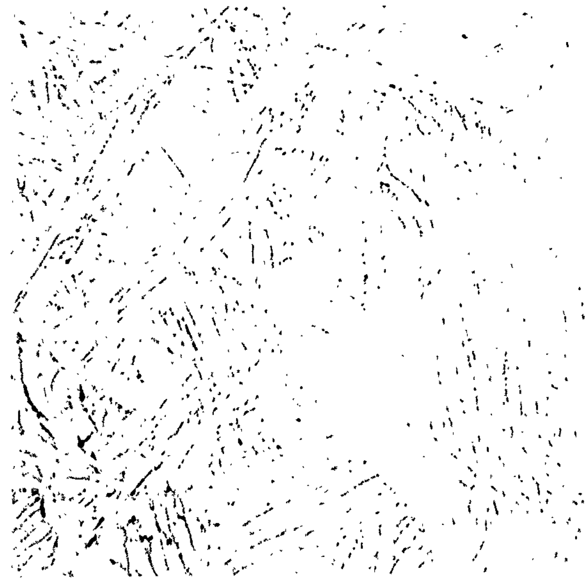


FIG. 52 - NOTCHED IMPACT STRENGTH VS TEST TEMPERATURE FOR Ti-6% Al(2) ALLOY.



Neg. No. 10586 X 200

Fig. 53

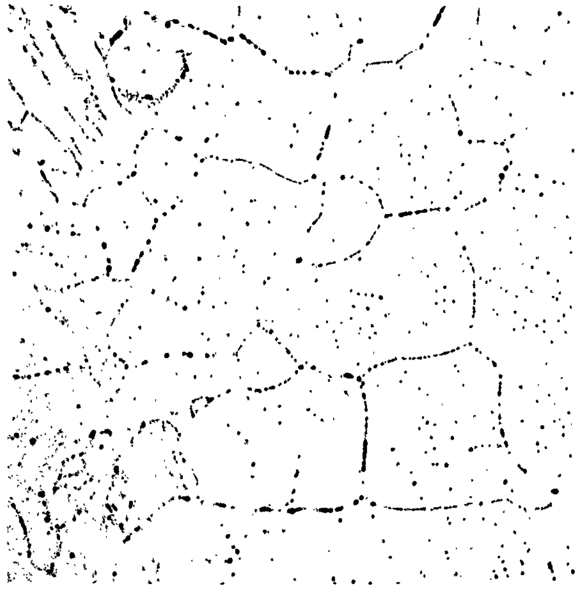
Ti-8Al (1J), annealed at 1200°F for 1 hour-air cooled, stress-rupture tested at 1020°F under 42,000 psi for 256 hours. α + precipitate.

Neg. 11519 X 1000

Fig. 54

Ti-8Al (2N), same thermal history as Fig. 43 plus 1470°F-2 hrs-WQ.

Etchant: 60 cc glycerine, 20 cc HNO_3, 20 cc HF.



Neg. No. 10657 X 250

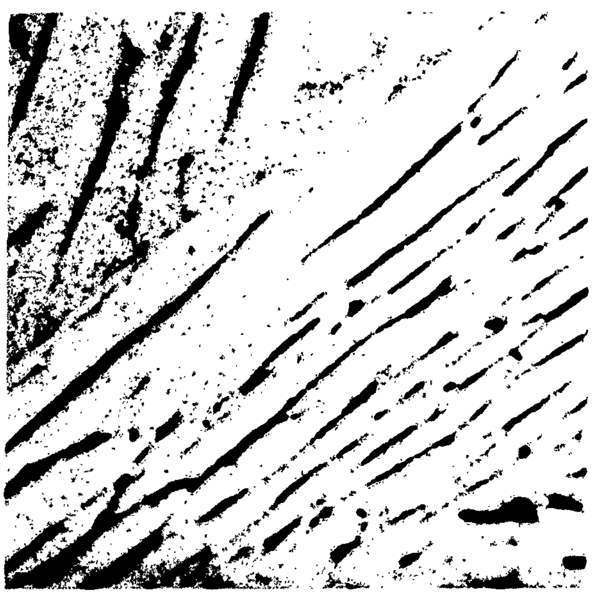
Fig. 55

Ti-8Al (2N), VA 1470°F-48 hrs-WQ, 1020°F-520 hrs-WQ.

Neg. No. 11520 X 1C

Fig. 56

Ti-8Al (2N), same thermal history Fig. 55 plus 1470°F-2 hrs-WQ.



Neg. No. 11514 X 1000

Fig. 57

Ti-8Al (Iod), same thermal history as Fig. 45 plus 1470°F-3 hrs-WQ.

Neg. No. 12103 X 15

Fig. 58

Ti-7Al (Iod), cold pressed 70%, 1470°F-180 hrs-WQ.

Etchant: 60 cc glycerine, 20 cc HNO_3, 20 cc HF.

Several attempts to make identification of the precipitate phase by X-ray and electron diffraction were unsuccessful. Finally, a Debye pattern of a powder sample of the 8Al specimen stress aged at 750°F for 1000 hours was obtained which gave unequivocal evidence of the presence of a second phase. Exposure conditions under molybdenum K_α radiation were as follows: 50 KV, 12 ma and 40 hours. Lines not belonging to α titanium are recorded in Table 23. Included in the table are the lines for the TiAl phase at 38% aluminum. Comparison of the two sets of lines strongly indicates that the second phase is TiAl (γ). Since ductility was recovered by heating embrittled specimens at 1470°F, a powder sample of stress-aged specimen was annealed at 1470°F for 1 hour and water quenched. The Debye pattern of this powder showed that the second phase did not go back into solution. This finding is consistant with the microstructural observations. The TiAl phase is known to be brittle.(6) However, the facts suggest that the embrittlement is due to coherency of pre-precipitation particles and that fully precipitated particles in the amount observed in this work are not harmful to ductility.

Further confirmation of the identity of the precipitate was obtained with a sample of 7Al alloy which was cold pressed 70% and subsequently annealed at 1470°F for 180 hours. The microstructure of this sample was clearly two phase, as shown by Figure 58. The Debye pattern of a powder sample of this material is also included in Table 23. As in the case of the 8Al specimen, the agreement with the TiAl (γ) pattern is rather good.

These findings show that the Ti-Al system as reported in the literature(7) is in error with respect to the extent of aluminum solubility in the α phase.

IV. SUMMARY AND CONCLUSIONS

The alloys Ti-5Al-2Ag and Ti-5Al-7Ag exhibited fairly typical age hardening response over the temperature range 800° to 1020°F. The 5Al-2Ag and 5Al-7Ag alloys have poorer creep-rupture properties at 800°F than the 6Al binary alloy.

Complexing the α phase of the 6Al-3Mo alloy with additions of tin and zirconium increased tensile strength with no sacrifice of room temperature ductility. Complexing the β phase by substitution of chromium, manganese and vanadium also effected an increase in strength with no impairment of room temperature ductility with the exception of the 6Al-2Mo-1Mn modification. Vacuum annealing this alloy restored ductility comparable to the base material. The 6Al-3Cu, α + Ti_2Cu alloy was exceptionally strong at elevated temperatures but had very poor or good room temperature ductility depending upon microstructure. Widmanstätten microstructures were brittle or of low ductility, while equiaxed structures were of relatively high ductility.

On a stress-rupture basis the α complexed alloys were slightly stronger at 1020°F than the 6Al-3Mo alloy. The creep-rupture properties of the β complexed alloys were significantly inferior to the base material at both test temperatures.

TABLE 23

TITANIUM-ALUMINUM ALLOYS

DEBYE LINES OTHER THAN ALPHA TITANIUM

7Al P-70* 1470°F-180 hrs		8Al* 750°F-1006 hrs 30,000 psi		TiAl 38Al 1290°F-330 hrs
I	d	I	d	
4 (?)	3.31	3	3.31	
2	2.87	4	2.83	2.81
1	2.14	1	2.10	2.08
1	2.02			
		<<1 (?)	1.93	1.99
		1	1.818	1.804
2	1.652	1	1.638	1.648
<<1 (?)	1.571	<1	1.536	
<<1 (?)	1.519			
		1	1.398	1.408
		1	1.384	
1	1.298			
1	1.279	2	1.271	1.257

* Mo radiation, 50 KV, 12 ma, 40 hours.

Considering creep resistance, the 6Al-3Cu and the α complexed alloys were comparable at 800°F and were markedly superior to the 6Al-3Mo alloy. At 1020°F the creep resistance of the 6Al-3Cu alloy was comparable to the 6Al-3Mo alloy, whereas the α complexed alloys were still decidedly better.

The 6Al-3Mo alloy was stable after creep exposure in agreement with previous findings for other Ti-Al-Mo alloys. Stability tests of the other experimental forging alloys gave variable results. In some instances exposure to stress at elevated temperatures produced severe loss of room temperature ductility, while in other cases for the same alloy comparable exposure conditions produced little or no property change.

Modification of the 6Al-4V sheet alloy by substitution of molybdenum for part of the vanadium indicated that the 6Al-2Mo-2V alloy was best of the three modifications both on a tensile strength and creep-rupture comparison. The 6Al-2Mo-2V alloy was appreciably weaker than a 6Al-4V alloy tested in bar form, but the great difference in strength is felt to be due in large measure to specimen geometry.

All three sheet modifications were indicated to be stable on the basis of limited stress-aging experiments.

The Ti-7Al-3Mo alloy had been found to have variable ductility depending upon hot working history as given the heat treatment: 1610°F-24 hours-air cool, 1020°F-24 hours-air cool. However, as heat treated: 1560°F-4 hours-air cool, 1020°F-24 hours-air cool, ductility was uniformly high. Vacuum annealing to remove hydrogen combined with the 1560°F-4 hours anneal resulted in further small improvement in ductility.

The 8Al alloy made with iodide titanium and vacuum annealed was found to embrittle in the same manner as the sponge-base alloy. A microstructural constituent is definitely associated with the embrittlement. The microstructural constituent, which was developed by long anneals at 1020° and 750°F, was not eliminated by reheating to 1470°F and quenching. However, electrical resistivity, which had been found to decrease when specimens were annealed in the temperature range of embrittlement, returned to the high value associated with the ductile state upon reheating to 1470°F and quenching. Also, tensile properties including ductility were completely reversible.

The microconstituent developed in the 8Al alloy was also developed in the 6, 10 and 16Al alloys. The 6Al alloy was ductile at room temperature (85°F) in all heat treated conditions. However, specimens annealed at 1020°F for 48 hours showed considerable loss in ductility when tested below room temperature. The 6Al alloy quenched from 1470°F also showed ductility loss when tested below room temperature, but the loss was considerably less than for the aged at 1020°F condition. While tensile properties of the 6Al alloy at room temperature were unaffected by aging at 1020°F, Charpy V-notch impact strength was decreased by a factor of two.

The 10Al alloy was brittle for all heat treatments. However, fracture stress was higher the higher the temperature from which the specimens were quenched. Highest quenching temperature was 2010°F. The lowest test temperature at which 10Al specimens aged at 1020°F showed some ductility was 480°F.

X-ray diffraction evidence indicates that the embrittling precipitate is the TiAl (γ) phase of the Ti-Al system.

The general results show that the Ti-Al system as reported in the literature is in error.

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APPENDIX

SOME OBSERVATIONS ON MICROSTRUCTURE

Five of the 8Al (Iod) tensile specimens for which data are given in Table 21 were sectioned longitudinally and examined metallographically. A common feature of the microstructures near the fracture surface was the presence of slip traces. Figure 59 shows slip traces of a specimen quenched from 1560°F and having 30% tensile reduction in area. Figure 60 shows slip traces of a specimen air cooled from 1650°F and then aged for 48 hours at 1020°F. This specimen possessed 3% RA. Slip traces such as these were also found in one of three fractured tensile specimens of 10Al alloy. The specimen had been quenched in iced brine from 1830°F. Slip traces were not observed in 6Al specimens fractured at room temperature.

The occurrence of slip traces was not limited to specimens which were plastically deformed. They were also observed in 8Al (iodide) specimens having the following thermal histories subsequent to forging:

(1) 1470°F-1 hr-WQ, 1290°F-1 hr-WQ

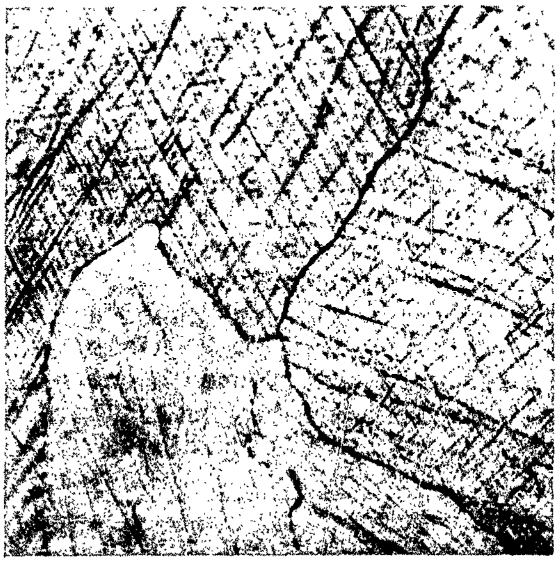
(2) 1470°F-1 hr-WQ, 1110°F-2 hrs-WQ

(3) 1470°F-1 hr-WQ, 1020°F-48 hrs-WQ

These specimens contained approximately 100 ppm hydrogen. Electronmicrographs of the specimen shown in Figure 60 at 11,000 and 22,000 X show the slip traces to be bands approximately 3×10^{-5} cm wide. The matrix was shown to be uniformly pitted with the bands considerably less pitted.

Ordinarily, under circumstances such as these, slip traces are not observed. Their presence indicates that something happens either during slip, or subsequently, during the preparation for metallographic examination which produces the etching effect. The slip traces present the possibility that precipitation occurs during slip.

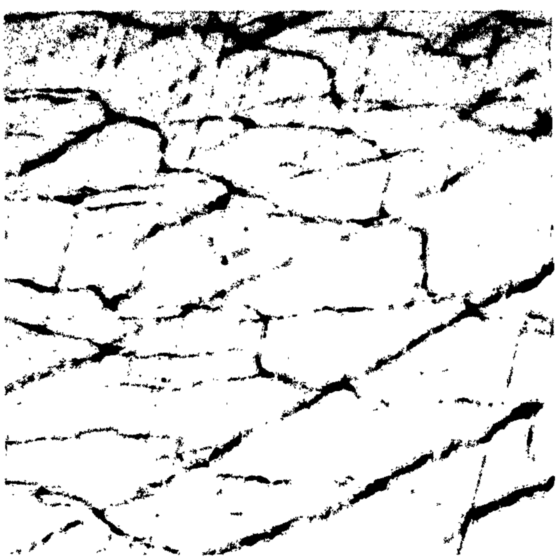
Another phenomenon observed, but to a more limited extent, was internal cracking. The cracks were transgranular. A specimen of 8Al alloy containing 100 ppm hydrogen was given the following heat treatment: 1560°F-6 hrs-WQ, 1020°F-500 hrs-WQ. The microstructure of the specimen showed copious precipitate in the grain and subgrain boundaries. The specimen was cut in half, and one half was solution heat treated at 1470°F for 2 hours and water quenched. The solution treated half of the specimen was found to contain numerous cracks (see Figure 61), while the other half was devoid of cracks. The mechanism of the formation of these cracks is unknown.



Neg. No. 11535 X 1000

Fig. 59

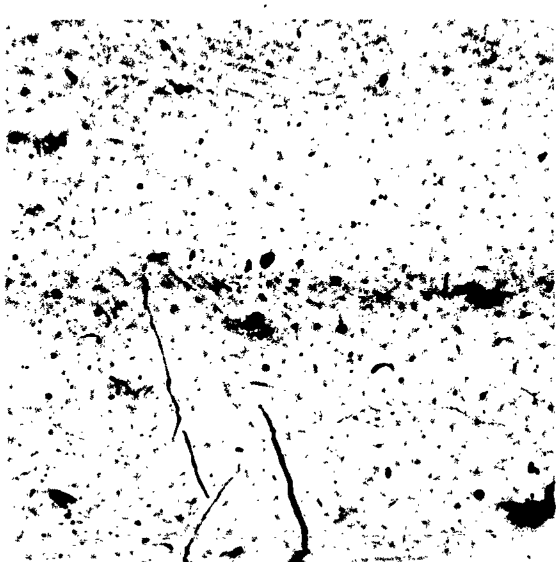
Ti-8Al (Iod), VA 1560°F-6 hrs-WQ. Fractured tensile specimen. RA = 32%.



Neg. No. 11431 X 2000

Fig. 60

Ti-8Al (Iod), VA 1650°F-96 hrs-AC, 1020°F-48 hrs-WQ. Fractured tensile specimen. RA = 3%.



Neg. No. 11518 X 250

Fig. 61

Ti-8Al (2N), 1560°F-6 hrs-WQ, 1020°F-500 hrs-WQ, 1470°F-2 hrs-WQ.

Etchant: 60 cc glycerine, 20 cc HNO_3, 20 cc HF.